BRITISH WELDING JOURNAL

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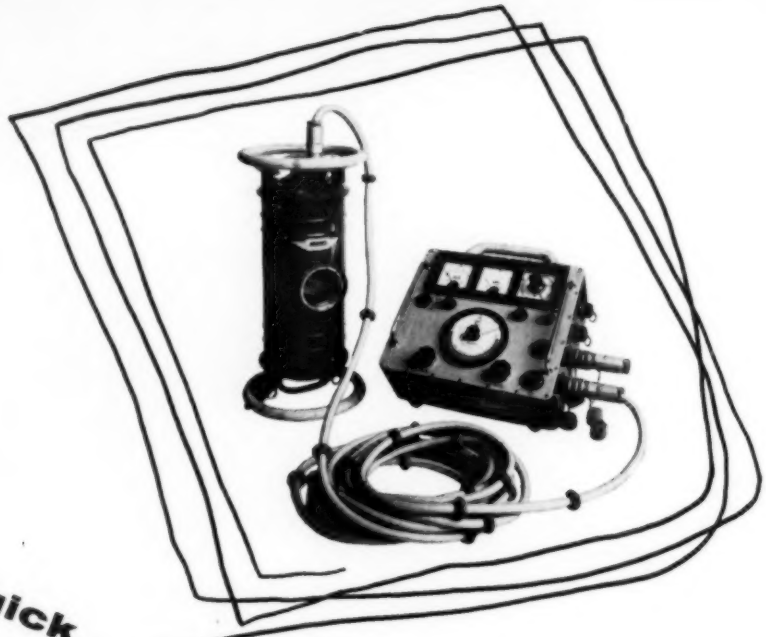

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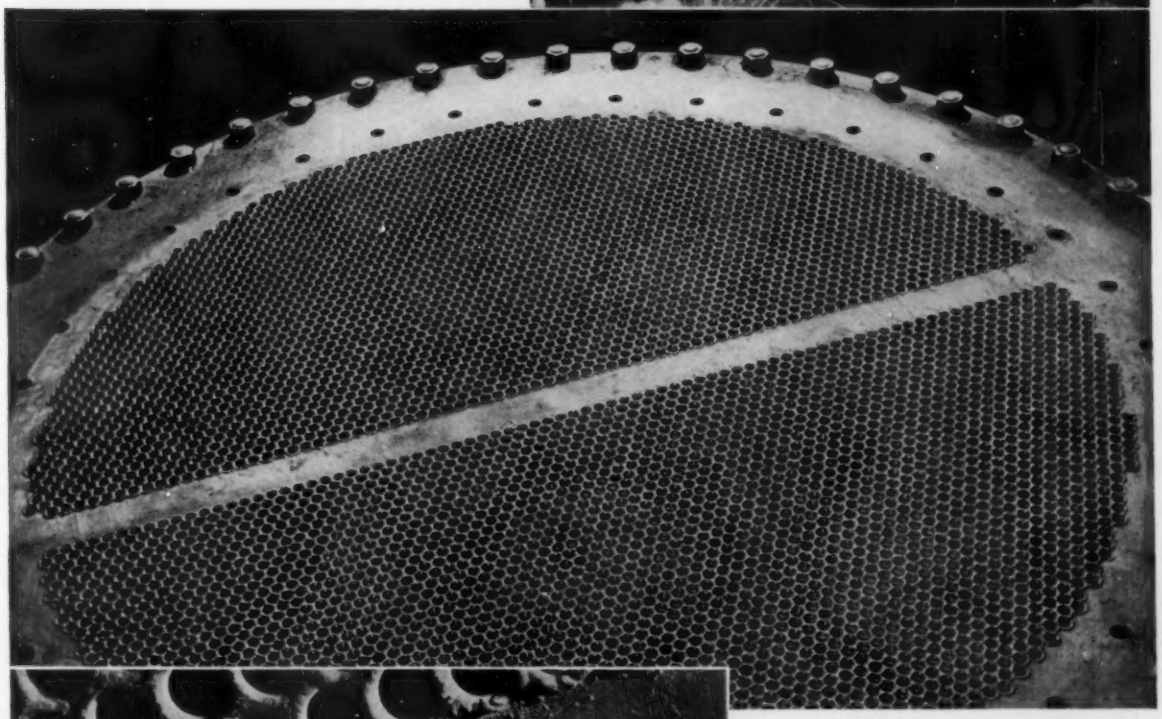
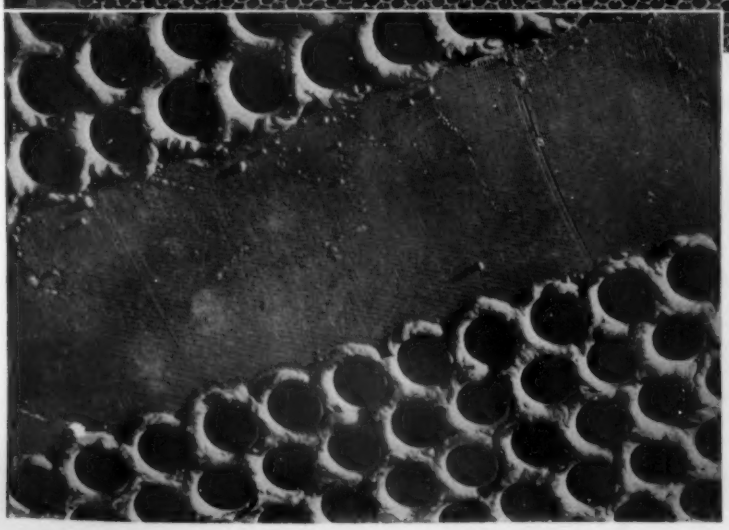
Inside front cover

Dexterity in WELDING





Tube plate for a Heat Exchanger, incorporating 5,518 tubes ¾ in. dia., and involving 1,078 footage of welding.

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WELDING OF NON-FERROUS METALS

Welded evaporator shell fabricated from I.C.I. copper sheet by Messrs. Caird and Rayner Ltd., London.

TECHNICAL NOTES: NO. 2:

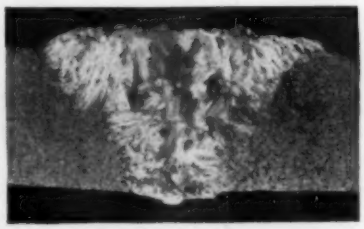
Welding Copper

The difficulties previously associated with welding copper have been largely overcome by the advent of inert-gas arc welding processes. The need for post-welding treatment is obviated, and joints with mechanical properties similar to that of the parent metal can be made by argon- or nitrogen-shielded tungsten arc welding or by argon-shielded metal arc welding.

For these processes, I.C.I. Metals Division has developed special filler rods and wires—'Argofil' for argon arc welding and 'Nitrofil' for nitrogen arc welding.

Imperial Chemical Industries Limited, Metals Division, produces large tonnages of non-ferrous metals and alloys for brazed and welded assemblies, and a wide range of rods and wires for brazing and welding.

The Company's Research Department has for many years carried out development work on the joining of metals. The experience gained is freely available to all interested in this subject.



Macrograph illustrating quality of weld which is easily obtained in I.C.I. phosphorus-deoxidised copper, using the argon-shielded metal arc process and 'Argofil' wire.

ICI METALS DIVISION

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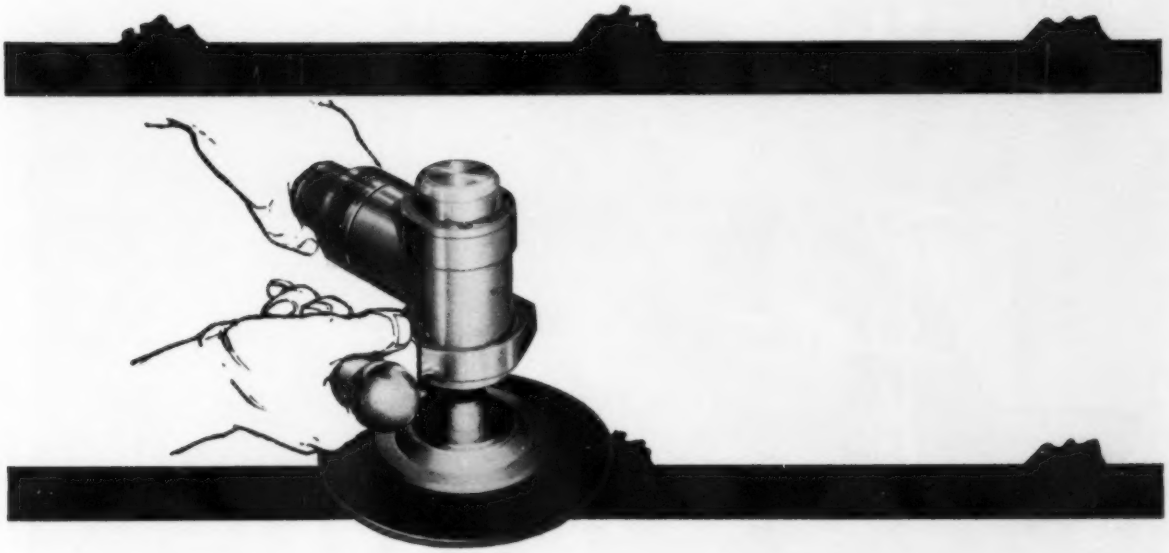
M.13

Taking the rough with the smooth...



CARBOFLEX wheels mean more speed, more safety and more economy in weld-dressing operations of all kinds. The resilience of CARBOFLEX depressed-centre wheels gives maximum tolerance to careless handling or unavoidable shocks: speeds up to 16,000 s.f.p.m. may safely be used. There is no more versatile tool. An easily chamfered edge cuts away excess weld almost fiercely, yet leaves a smooth undamaged surface. CARBOFLEX depressed-centre wheels have a multitude of uses including de-scaling and de-burring, smoothing flame-cut edges, cutting sheet metal, grooving seams for welding, bevelling sheet metal, and cleaning up castings. They are most useful tools in any welding shop.

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Zeta, with inset showing BTH 'clamping' ignitron.

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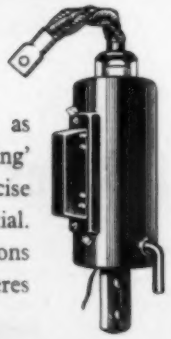
IGNITRONS



IN ZETA, BTH Ignitrons are employed as electronic switches for 'discharge' and 'clamping' duties, and perform vital tasks where precise timing and complete reliability are essential. Specially developed for the project, these ignitrons carry peak currents in excess of 20,000 amperes in the 25,000-volt power circuit.

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EUTECTIC Welding News

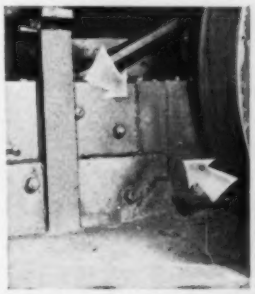
PUBLISHED BY EUTECTIC WELDING ALLOYS COMPANY LIMITED, FELTHAM, MDDX.

THE HISTORY OF "LOW HEAT INPUT" WELDING

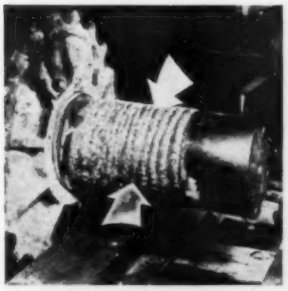
In 1904 the late J. P. H. Wasserman, a pioneer Swiss metallurgist, first observed the phenomenon of surface alloying during brazing operations. His son, Rene D. Wasserman, continued research and perfected the new "surface alloying" metal joining and repair process. Today, advanced Eutectic "Low Temperature Welding Alloys" minimize or eliminate the distortion, warping, embrittlement, cracking and other undesirable metallurgical changes in base metals which can occur with conventional high heat materials. Throughout Great Britain, a large force of factory-trained technical representatives is bringing to industry the benefits of Eutectic's "Low Heat Input" Process. Take advantage of this free service by completing the coupon below and posting it to us today.

XYRON 2-24 SAVES COSTLY SHUT-DOWN

Fractures suddenly developed in the cast iron base of this cable hoist. Faced with a complete stoppage of production and a lengthy repair job, the manufacturer used Eutectic Xyron 2-24. With this high quality, "Low Amp" electrode, the fractures were repaired in position. Dismantling and preheat were not necessary. Xyron 2-24 enables the heaviest and most complicated cast iron jobs to be "cold" welded in all positions. Preheat and after-cooling necessary with conventional material are eliminated. Xyron 2-24 operates at low amperage A.C. or D.C., gives an extremely dense, smooth deposit, with a tensile strength of 13½ tons p.s.i.



9 TON SHAFT OVERLAYED BY EUTECROD 185 IN 8 HOURS



The bearing end of this nine-ton, 23 ft. stoker drive shaft was found to be badly worn, after 21 years of continuous operation. A large public utility company thought repair would be lengthy and expensive until Eutectic's Technical Representatives suggested EutecRod 185, a patented BronzoChrom overlay with the highest resistance to frictional wear. EutecRod 185 repaired the shaft in only eight hours, with great savings in material and labour.

The overlays were applied without raising the shaft above dull red in colour, and EutecRod 185 bonded without fusion of the base metal at between 540° and 650°C.

EutecRod 185 whilst harder and tougher than conventional overlays is fully machinable. Characteristics include high ductility, low coefficient of friction and excellent resistance to corrosion with highest resistance to wear and impact.

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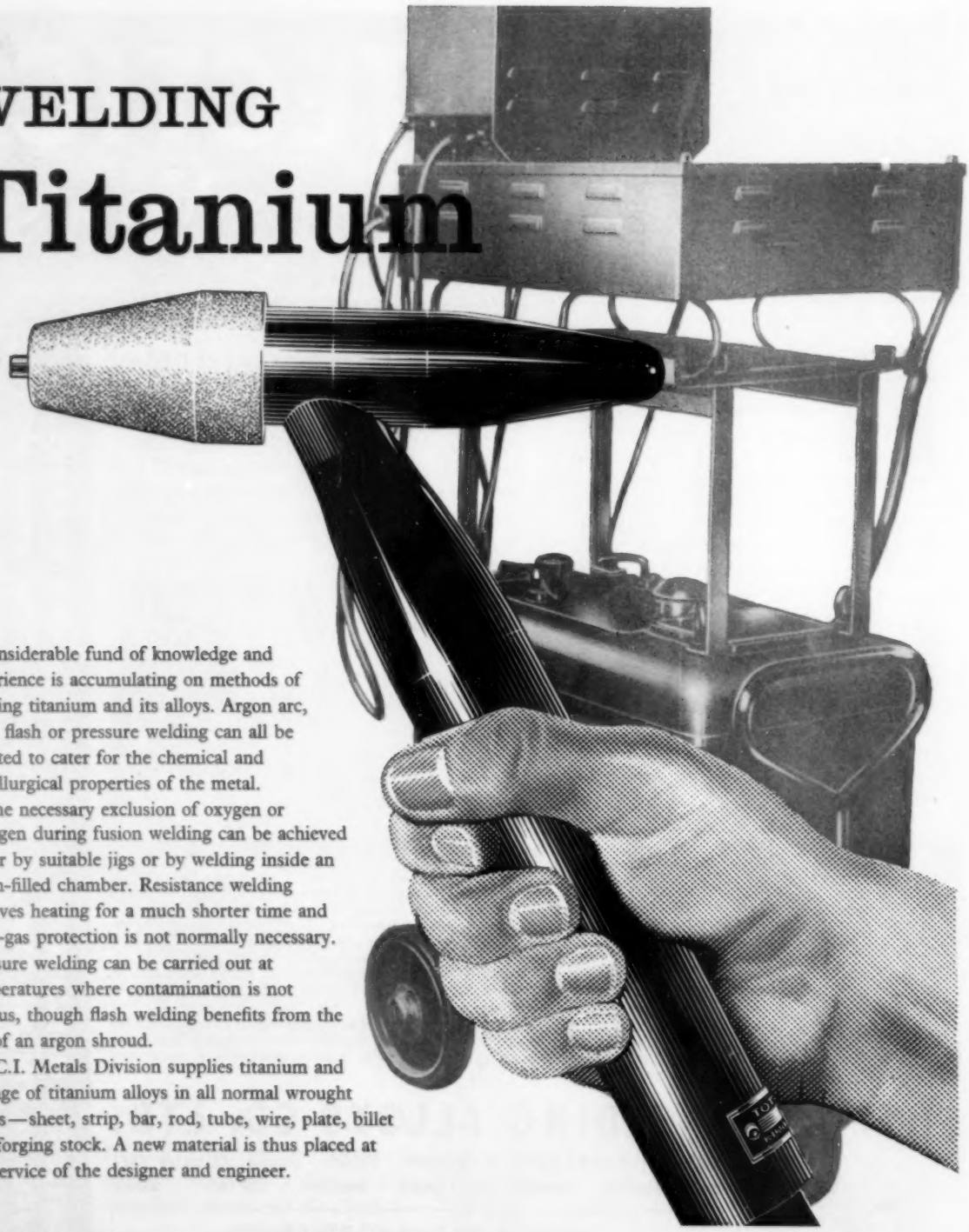
EUTECTIC 'LOW TEMPERATURE WELDING ALLOYS' FORMULATED FOR 'LOW HEAT INPUT' BONDING OF ALL METALS INCREASE SAVINGS BY SALVAGE AND REPAIR, SOLVE PRODUCTION PROBLEMS

WELDING Titanium



A considerable fund of knowledge and experience is accumulating on methods of welding titanium and its alloys. Argon arc, spot, flash or pressure welding can all be adapted to cater for the chemical and metallurgical properties of the metal.

The necessary exclusion of oxygen or nitrogen during fusion welding can be achieved either by suitable jigs or by welding inside an argon-filled chamber. Resistance welding involves heating for a much shorter time and inert-gas protection is not normally necessary. Pressure welding can be carried out at temperatures where contamination is not serious, though flash welding benefits from the use of an argon shroud.

I.C.I. Metals Division supplies titanium and a range of titanium alloys in all normal wrought forms—sheet, strip, bar, rod, tube, wire, plate, billet and forging stock. A new material is thus placed at the service of the designer and engineer.

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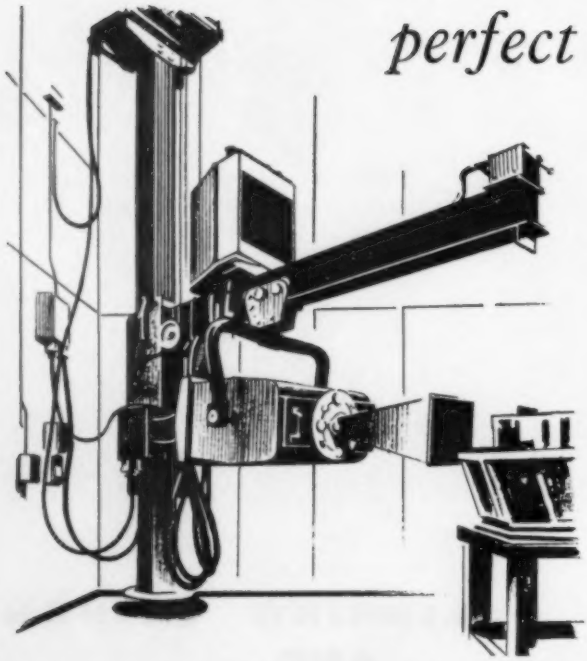

Few requirements in modern engineering are more exacting than those imposed on long-range aircraft such as the Bristol Britannia, with their arduous, non-stop operational schedules. Welds and castings in the aircraft are therefore subjected to rigorous tests in production by radiographic methods. For this vital scrutiny, Newton Victor industrial X-ray units are used. The Raymax 250 has been specially designed for industrial radiography either indoors or on open constructional sites, and in every branch of engineering where perfect welds are essential. Its use for the Bristol Britannia is one further demonstration of its reliability.

The Newton Victor Raymax 250 positioned for examining an engine casting.

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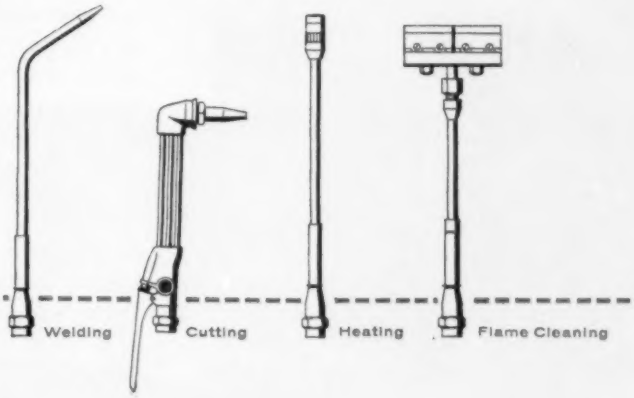

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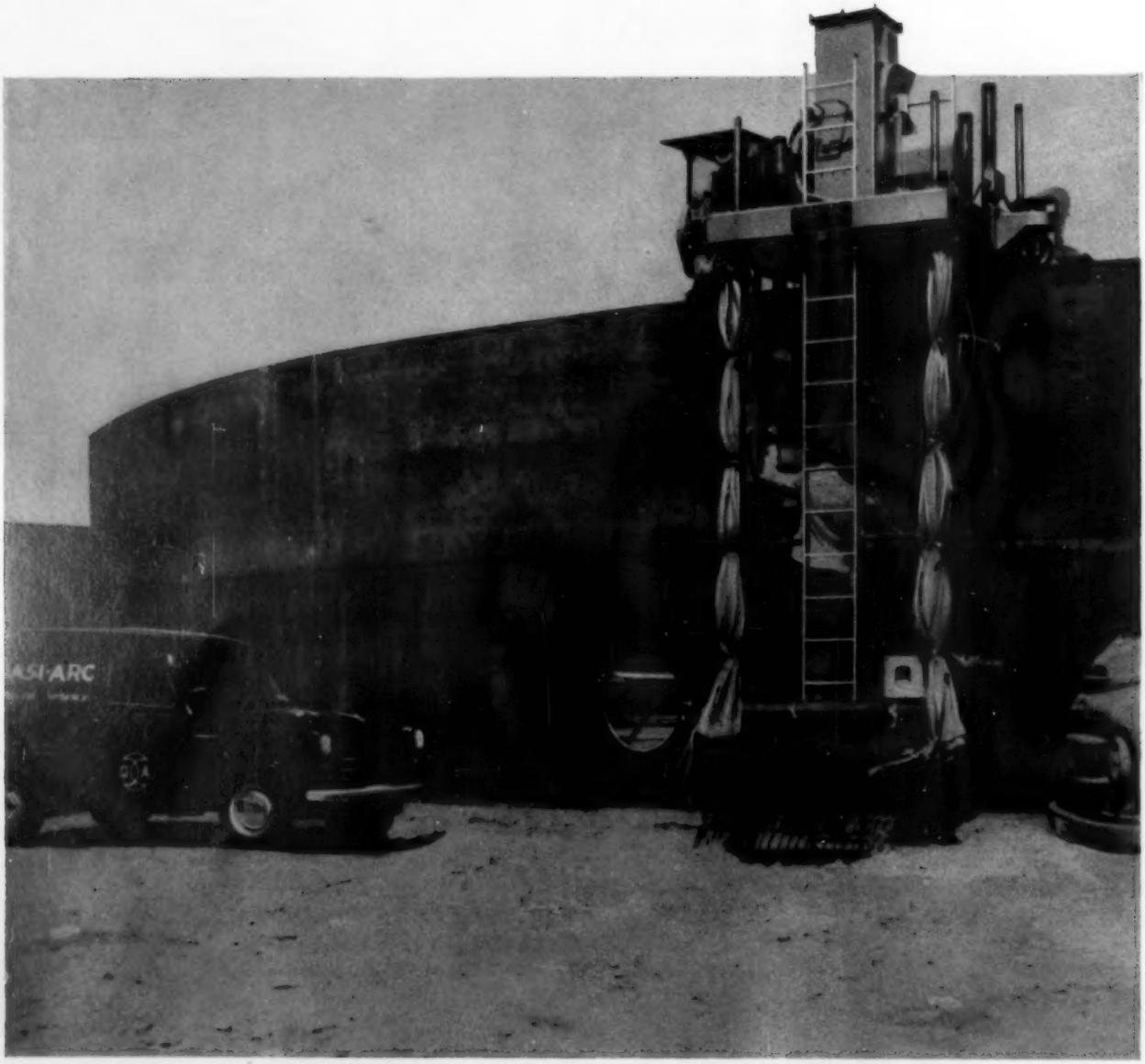
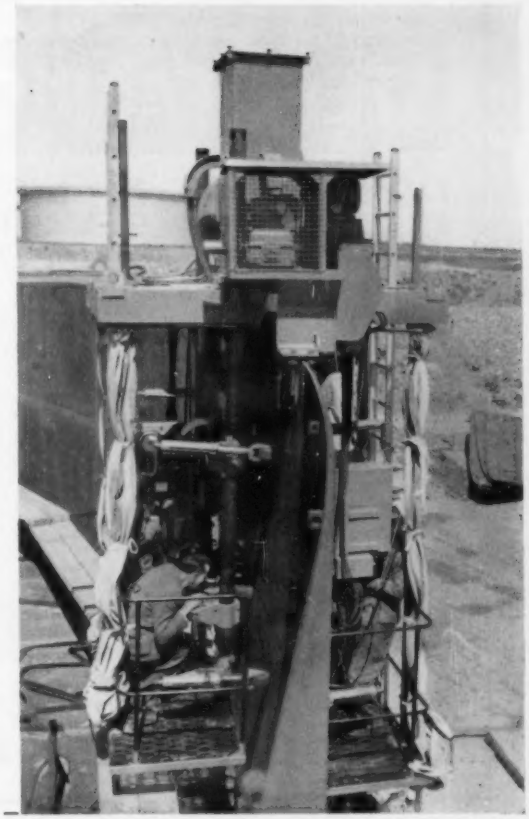
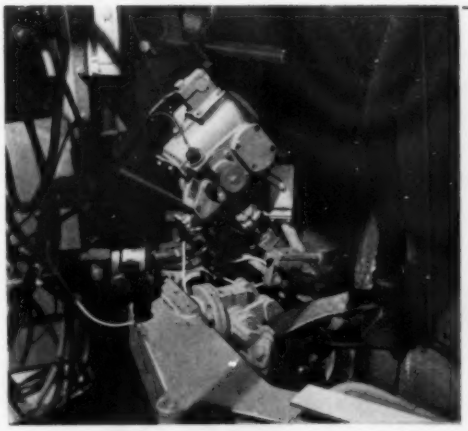


This is the Quasi-Arc–Pandjiris automatic girth welding equipment, specially developed for the high speed site welding of oil storage tanks.



THE QUASI-ARC—Pandjiris Girth Welder consists of two welding heads with control panels and platforms for two operators. These are mounted on a saddle-type carriage which straddles the top strake of the tank. The drive and idler wheels of the carriage run directly on the curved edge of the plate and guide rolls are provided to act as a pivoting device. Thus there is one welding head inside the vessel and one outside, with an operator on the platform of each. As the carriage moves at the selected welding speed round the perimeter of the tank, the internal and external seams are welded simultaneously and continuously, in the horizontal position.

Vertical adjustment for the welding heads allows for plate widths of 4 ft. to 9 ft. with additional local adjustment of 12 in. for accurate positioning of the welding heads. A spot-light guiding device is also fitted to enable the operator to guide the welding nozzle accurately along the joint. The carriage movement is remotely operated from the operators' control panels and speed indicators are fitted. Holders are provided for oxygen and acetylene cylinders and torches for pre-heating the joint ahead of the welding nozzle. Telephone sets for constant communication between the two operators are also provided.



The UNIONMELT welding composition in this installation is supported under the welding nozzles by an endless belt arrangement, tilted so that the composition presses against the joint. Support is thus provided for the molten weld metal so that welding currents up to 750 amps can be used. A continuous melt circulation system recovers the unfused composition and feeds it back to the hopper at the top of the machine; from there it falls by gravity back to the welding nozzles. As each strake is completed, removal and re-positioning of the equipment for the succeeding strake is simple and rapid.

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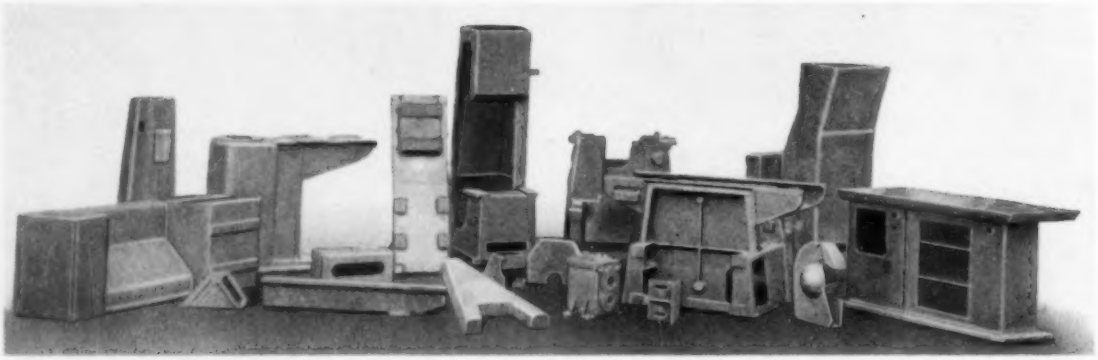

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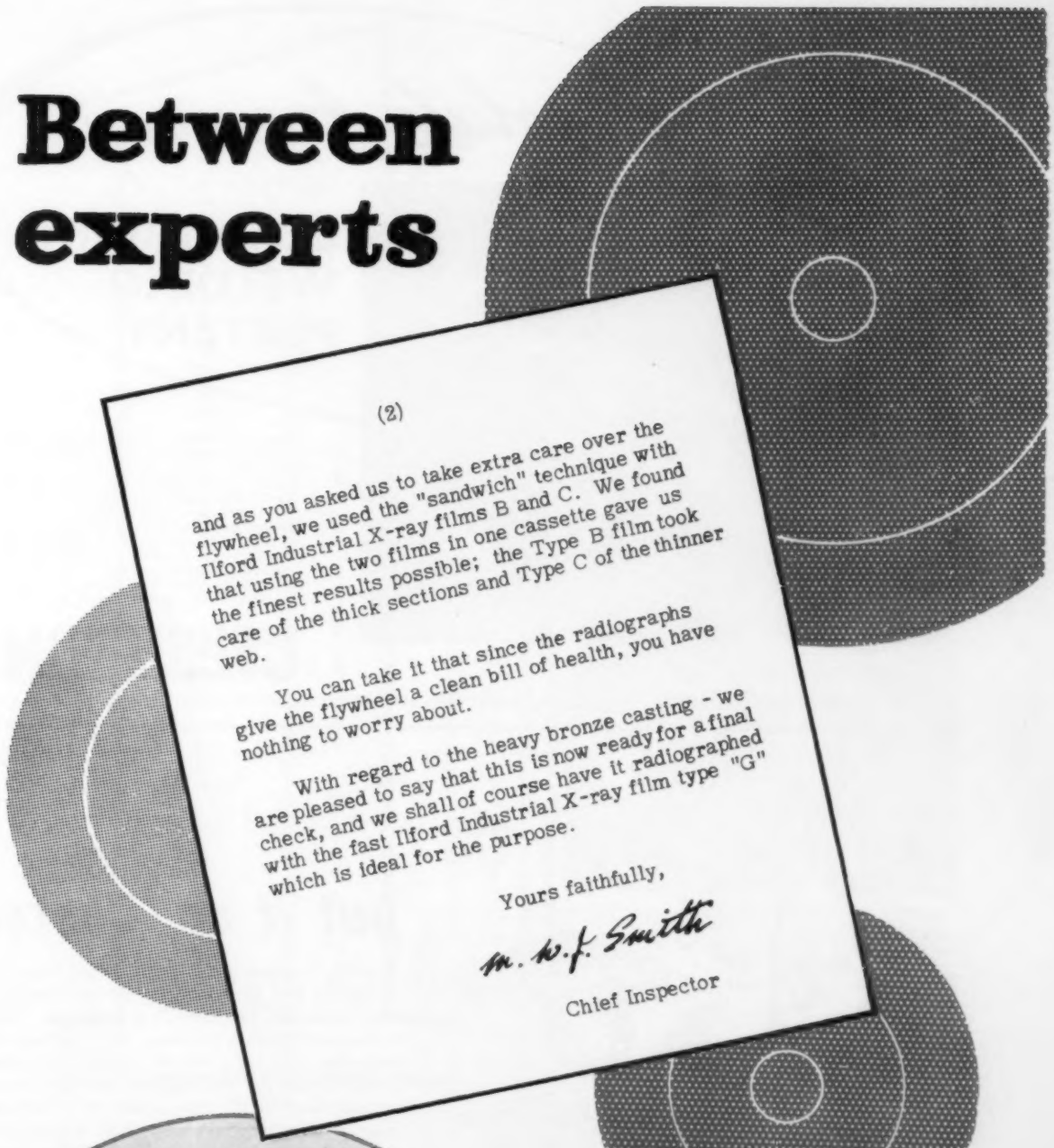

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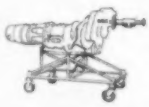

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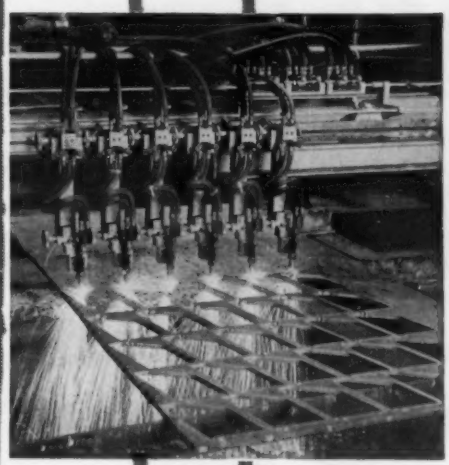

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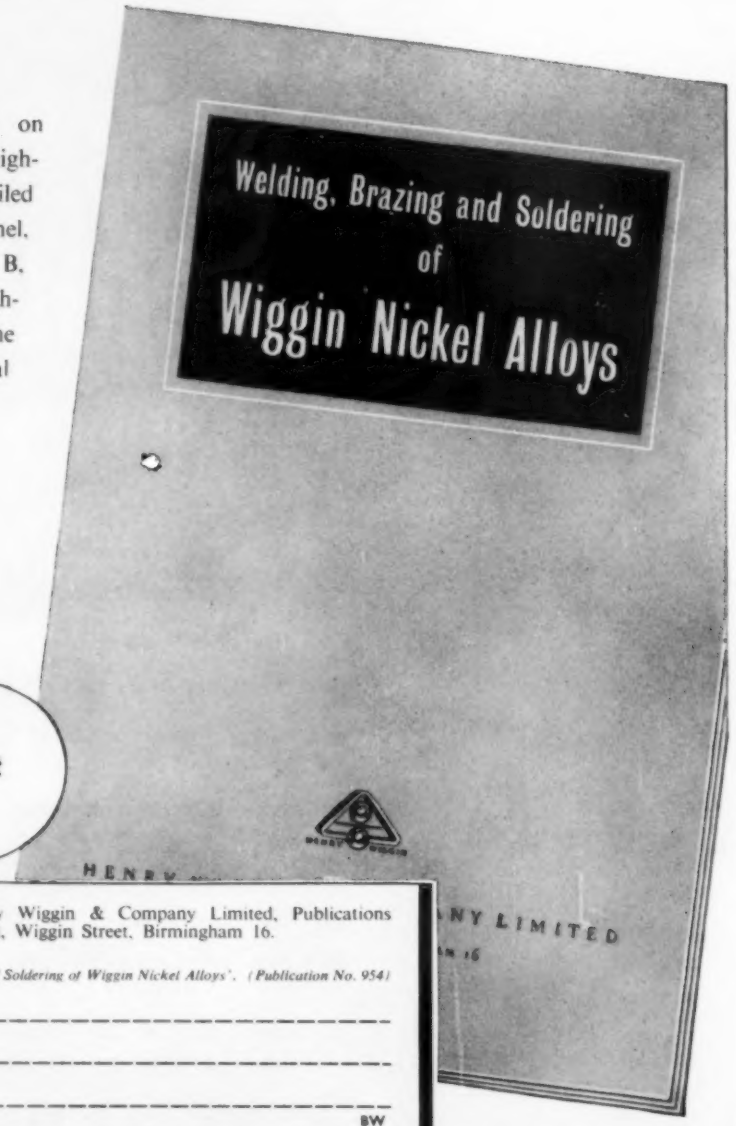

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British WELDING Journal

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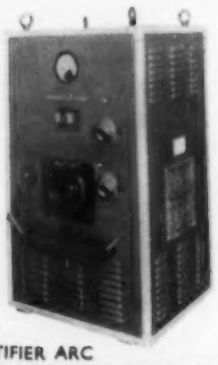

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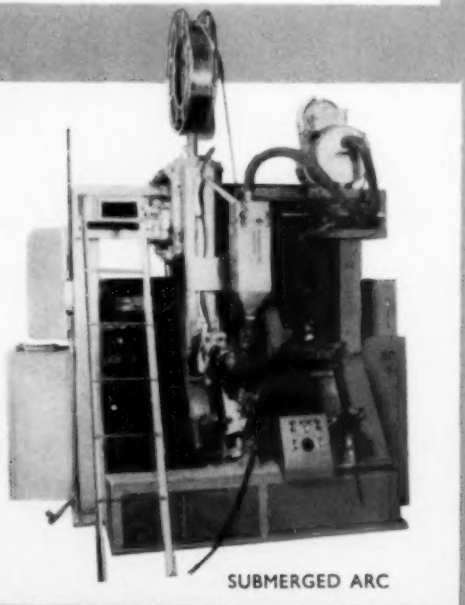

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Editorial

THE INTERNATIONAL INSTITUTE OF WELDING

Though this Journal has provided its readers with a generous selection from the reports and memoranda issued in increasing number by the International Institute of Welding over the past four or five years, the influence of that body is probably less widely appreciated and its organization less clearly understood than they deserve to be. The tenth anniversary of the foundation of the International Institute in May, 1948, prompts a brief survey of its development to the present time.

International co-operation in welding hardly existed before the foundation of I.I.W. It would be a considerable achievement in ten years if the I.I.W. had done no more than enlist the active participation of 25 countries. Yet it has done much more than that. Modelled on the International Standards Organization, and rejecting the example of those international bodies which at intervals of five or ten years hold a large congress and add another large volume of papers to the shelves, I.I.W. labours continuously through fifteen technical Commissions and more than thirty Sub-Commissions and Working Groups, covering in detail a wide area of technical study.

Four of the main Commissions are concerned with the welding processes—gas, electric-arc, resistance, and the special processes of more recent origin. Testing and Measurement, Weldability, Residual Stress and Stress Relieving, Pressure Vessels, Boilers and Pipelines, Fatigue Testing, and the Fundamentals of Design and Fabrication are the subjects studied by six more, and the remaining five are assigned to Standardization, Hygiene and Safety, Instruction, Documentation, and Terminology.

The extent and perhaps the value of the work accomplished has certainly been obscured by the mode of publication. So far, the I.I.W. has not published a journal of its own: its reports and memoranda are distributed for publication in the journals of its member-countries, where engineers can read them in their own language. Regular publications such as the *Bibliographical Bulletin* (a comprehensive quarterly of welding abstracts), the *Multilingual Collection of Terms*) (a glossary of welding terms in twelve languages), and the *Radiographs of Welds* are issued by member-societies on its behalf. In sum, over the ten years, these reports and memoranda total some hundreds of separate items, large or small, and the rate of output has been rising steadily each year. Subsidies granted by UNESCO through the Union of International Engineering Associations for the *Bibliographical Bulletin* and the *Multilingual Collection of Terms* show that the standard achieved commands respect.

The desire to join in the work of I.I.W. has led to the foundation of Welding societies and institutes in several countries where none previously existed: Italy, Spain, Yugoslavia are examples. In other countries it has prompted a number of existing bodies to set up a co-operative organization among themselves, to handle their relations with I.I.W. That is the situation in Britain, where a Joint Committee, now under the Chairmanship of Mr. A. Robert Jenkins, one of the present Vice-Presidents of I.I.W., was established in 1948 by the U.K. member societies.

It is this Joint Committee which appoints the U.K. representatives on the administrative and technical committees of I.I.W., briefs them before the annual assemblies, and assigns to each an appropriate advisory committee from among the committees appointed by the member societies.

The British Institute of Welding carries a substantial share of the responsibility for this international work. While the French Institute provides a Scientific and Technical Secretariat, the Secretary-General of I.I.W. from the beginning has been the Secretary of the Institute, and much of the Assistant Secretary's time has been given to international business. So too the Institute provides a secretariat for the Joint Committee in its task of organizing British participation in the I.I.W.

Substantial benefits are earned by these activities and are passed on to members through the publications, meetings, and library service of the Institute. More tangible benefits, it may be supposed, accrue to the British Engineering industry, if the growing demand for admission to I.I.W. assemblies means what it implies—an alertness to the importance of putting forward the British viewpoint with authority, and of showing that this country is in the forefront in the use of modern methods of metallic construction. The establishment of a Free Trade Area should increase the importance of these international contacts to British exporters.

Welding
TECHNOLOGY

Welding in Ship Repair

By N. G. Hanson

SYNOPSIS

In describing a few examples of repairs to shell plating and bulkheads the author refers to a semi-automatic method of cutting plates. This leaves an edge that is reasonably straight and chamfered ready for welding. It is considered that, if machines of the type developed were to be made available by the equipment manufacturers, a considerable saving of labour and time would result.

In this very brief paper it is not possible to give a comprehensive account of welding in the ship-repair industry. Nonetheless, it is thought that a description of some of the author's experiences in this field would be valuable.

In the repair of welded ships it has been the practice, and in many instances it probably still is the practice, to consider the repairs in relation to what has been done for many years on riveted ships. These are generally considered as being made up of many separate parts, assembled piece by piece and riveted together to form the completed structure. Construction has been orthodox, and new methods such as prefabrication have seldom been used.

The development of welding has, however, generated a new outlook, and the breaking down of a ship's structure into large sections for prefabrication is now accepted as standard practice in shipbuilding. In varying degrees the same methods have also been accepted for repairs on welded ships, particularly for oil-carrying vessels. A well-planned welding layout, better cranage, a prefabricating shed, and piped oxygen and acetylene gases are now essential features of a ship-repairing establishment.

The following brief examples of particular instances of repairs to welded vessels may show that there is a definite trend towards more effective methods of dealing with the problems involved, leading to more rapid completion and reservicing of vessels under repair.

SHELL PLATING

In the past, the renewal of a welded shell plate was a somewhat tedious job. The damaged plating was burned away with oxy-acetylene burners to leave plate edges in a state that depended largely upon the skill of the operators. After the damaged plate had been removed, machine caulkers faired up the edges as best they could and at the same time produced the necessary chamfers for welding. The result was invariably an edge that was far from straight—but as good as could be obtained under difficult circumstances. A template would then be prepared to the shape of the opening with its attendant defects, but in spite of every care the new plate would be a poor fit and machine caulkers would be needed to cut a little from here and there so as to obtain a reasonable fit.

Some years ago it occurred to the author, as it must have occurred to many others, that the solution to this problem lay simply in obtaining a straight-line cut, preferably already chamfered. Such a line would greatly reduce the work previously required of the machine caulkers and oxy-acetylene burners, and would also simplify templating and the final fitting of the new plates.

Experiments were carried out, with particular attention being paid to overhead burning so that damage to bottom plating could be dealt with by this method. A satisfactory means of obtaining a single chamfer in the overhead position was obtained. A machined aluminium track was used to carry and guide an

This paper was presented to members of the North Eastern (Tyneside) Branch of the Institute at a meeting on 5th December, 1957.

The author is with Swan, Hunter and Wigham Richardson, Ltd., Wallsend upon Tyne.

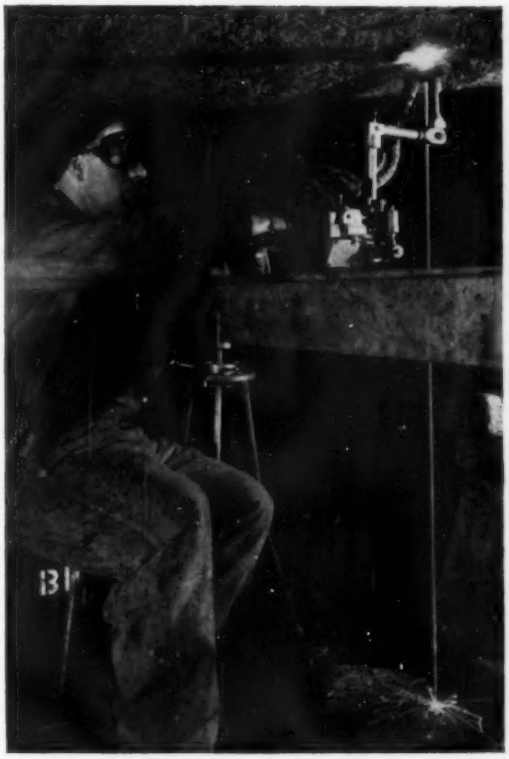

Fig. 1—Single-nozzle burner set up on track for cutting bottom-shell plating

electrically driven oxy-acetylene torch with a single nozzle. The track was supported on tripods with telescopic legs to allow for inequalities in the level of the dock bottom. The machine was a standard type with a modified support for the burner nozzle to suit it to overhead cutting (Fig. 1). Manual adjustment in the vertical and lateral directions was also provided. Before starting to cut the plates, it was necessary to descale both surfaces and to strike a chalk line for the guidance of the machine operator. A specimen single chamfer cut by this method is shown in Fig. 2. The only additional work required was a light chamfer with a machine caulker on the back face for the sealing runs. Although the finished line so cut was not exactly straight, it was a great improvement on previous results, and the labour costs were considerably reduced.

Fig. 2—A chamfered edge produced by the machine shown in Fig. 1

Fig. 3—The resulting straight cut from a double-nozzle burner

Initially, it was the practice to chamfer the 'inside' edge of the bottom shell plating so that welding could be carried out in the downhand position. More recently this has been reversed, so that the bevel is on the underside of the plating. This was done firstly because the vessel being repaired had carried a cargo of crude oil and welding over the deposits on the surface would have been difficult, and secondly because welding from the dock bottom provided more space than would be possible inside the vessel.

The further step of applying this method of cutting to side-shell and deck plating was equally successful. For side-shell plating repairs the machine track was supported on small brackets welded to the plating.

In subsequent trials during the repair of an all-welded tanker, an attempt was made to mechanize the process fully by producing a cut that needed no

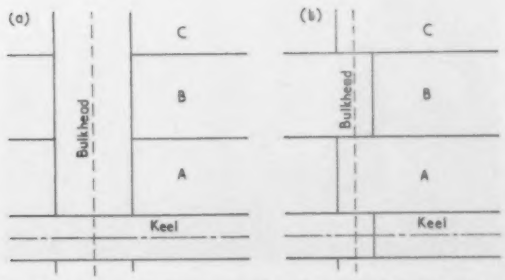

Fig. 4—Repair of bottom-shell plates in way of bulkhead: (*a*) original plates; (*b*) renewed plating with panel eliminated

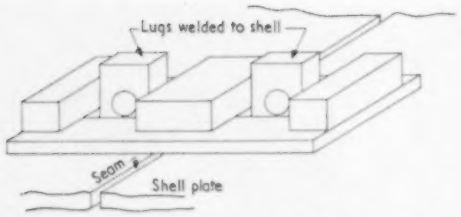

Fig. 5—Device for positioning and holding shell plates before welding

further work on it. A machine was used fitted with two burning nozzles (Fig. 3). At first a double bevel was attempted on a bottom plate, but because of difficulties in controlling the slag this was abandoned in favour of a vertical cut with a single bevel on the lower edge. The only additional work needed was the provision of a machine-caulked chamfer on the top edge.

It was found advisable to cut a plate seam in one length on a prepared setting of track rather than to do it piecemeal in lengths of, say, 12 ft at a time. Any unevenness of the plate surface was followed by manual adjustment of the nozzles. So far no satisfactory automatic method of maintaining a constant distance between nozzle and plate has been devised.

It is not always necessary in repairing a welded hull to replace the plating in its original form. For example, where the plating shown in Fig. 4*a* is to be renewed complete with internals, the panel in way of the bulkhead may be eliminated so that the renewed plating is as shown in Fig. 4*b*. This method allows the bottom-shell longitudinals to be welded directly to the plates before erection, so eliminating much fairing of the structure and simplifying the erection. Experience has shown that both owners and the classification societies are amenable to such modifications in design, which considerably reduce the work of the repairer.

A device for holding shell plating in position for welding is illustrated in Fig. 5. It eliminates the need for the many welded lugs, bolts, etc. normally used, and reduces the amount of fairing required.

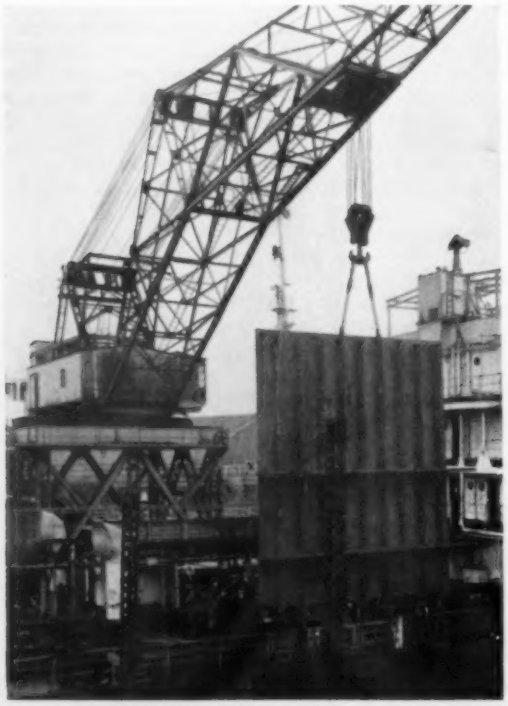

Fig. 6—Prefabricated bulkhead being lowered into position through opening provided by removal of deck strake

BULKHEADS

The repair of bulkheads presents a different problem from that of shell plating, because access for large sections is impossible. In some instances, however, complete units can be prefabricated to replace damaged sections. Figure 6 shows such a unit being lowered into position. Access was made available by the removal of the riveted deck plates over the fore-and-aft bulkhead, and the transverse plating width of the bulkhead cruciforms was governed by this opening.

In other positions, for example in way of the bridge space, fore-and-aft and transverse bulkheads were constructed as smaller units and were erected piecemeal, as were all the wing-tank bulkheads.

On all-welded tankers needing repairs to odd bulkhead plates, stringers, centre girders, etc., the difficulties of access arise, for the removal of a deck strake cannot always be justified. In these instances the plate widths of the new material must be made to suit the hatch openings, resulting in increased amounts of welding and fairing, and adding to the handling difficulties. Particularly for stringers, assembly and welding *in situ* is costly.

The repairer of welded tankers must often wish that shipbuilders would incorporate a riveted strake of plating over the fore and aft bulkhead so as to facilitate access to the cargo tanks.

Where odd bulkhead plates are to be renewed, it has been found advantageous to lapweld the seams rather than to use butt welds. This eliminates a considerable amount of fairing and simplifies erection. It also allows plate edges to be joggled to avoid the necessity of fitting packings on stiffeners and stringers—a special advantage on bulkheads of corrugated construction.

CONCLUSION

From the very few examples described it is evident that progress in welded ship repair depends largely on two factors:

(i) Improved methods of plate cutting and edge preparation *in situ*
(ii) An extension of the principle of prefabrication.

Although manufacturers of equipment in the U.K. have been aware for some time of the methods developed in the author's yards, it is disappointing to find that, so far, the production of the special cutting machines described is not yet contemplated. Yet progress in the repair industry, as in most others, depends primarily on the technological improvements of the machines used.

Welding

RESEARCH

The Welding of High-strength Heat-treated Aluminium Alloys

By P. T. Houldcroft, B.Sc.

SYNOPSIS

THE high-strength heat-treated aluminium alloys have not been recommended for welding in the past because of a tendency to cracking and low-strength joints lacking in ductility. It appeared that these difficulties might be overcome by the development of a filler material designed to give resistance to cracking and sufficient strength to ensure failure in tension in the heat-affected zone. Using a new type of cracking test, also ageing and tensile tests on welded joints, a filler material containing 7·0% Cu, 3% Si, and 18% Mg has been developed which has given satisfactory results over a wide range of dilutions.

Introduction

HIGH-STRENGTH aluminium alloys of the Al–Cu–Mg (H.14, H.15, L70–L73) and Al–Zn–Mg (D.T.D. 5050 and 5060) types have many useful properties but they are generally considered to be unweldable. This is unfortunate because a satisfactory technique for welding the alloys would be useful in certain structural work and in the replacement of castings and forgings by welded fabrications. Several years ago the B.W.R.A. began investigations to discover, in the first place, what the difficulties were and, secondly, to devise suitable techniques for overcoming them. The present report is a summary of this work and a statement of the present position. Although the investigations cannot be considered complete, sufficient is now known about the difficulties to justify the application of welding to the high-strength alloys in selected applications.

Report *LM* 8/26/58 of the Association issued to members in February, 1958.

The author is with B.W.R.A., Abington, Cambs.

THE PROBLEM

Many attempts have been made to weld the high-strength alloys[1] but in general the results have been disappointing with joint efficiencies of about 50% and severe trouble from weld cracking. Among the best results obtained in the 'as-welded' condition were those obtained with the use of an Al–9% Mg filler[2] and parent-metal filler.[3] Several investigators have stated that the heat-treatable alloys must be completely re-heat treated after welding to restore their properties. Even with complete re-heat treatment, however, joint efficiencies were frequently well below 90%. Rather more encouraging results were obtained in a series of *ad hoc* tests carried out at B.W.R.A. in material of various thicknesses welded with a number of filler materials of types readily available.[5] It was observed that the highest strengths were obtained with parent-metal filler, although Al–6% Mg alloy gave good results in both as-welded and re-heat treated joints. The Al–5% Si alloy was found to be quite unsuitable for welding Al–Zn–Mg alloys since it rendered the weld bead irresponsive to heat treatment. Elongation measurements (on 2-in. gauge length) indicated a poor overall ductility of less than 5%, but perhaps the most interesting observation was that the fracture in tensile testing was always in the weld. The

situation regarding the welding of high-strength heat-treated alloys at this time was that they were not welded commercially because:

(i) Severe cracking was encountered, especially if parent-metal filler was employed
(ii) Joint strengths were low unless parent-metal filler was used and the joints were subsequently re-heat treated
(iii) Joints were 'brittle' with poor elongation properties
(iv) It was thought that the heat of welding would destroy the heat-treated condition completely, making re-heat treatment a fundamental necessity. This of course was a discouragement in applications where it was a physical impossibility to solution heat-treat the completed component
(v) It was also thought that the heat of welding would result in serious overheating alongside the weld, producing a brittle condition locally which could not be corrected
(vi) There was evidence that stress corrosion would be a problem in even mildly corrosive environments.

THE APPROACH

It is interesting to examine these six reasons in greater detail. There seems to be little doubt that the fears regarding stress corrosion were well founded. For the applications envisaged, however, the welded assembly could be well protected. There was also the possibility of using clad material and protecting the weld metal locally. It was suspected that the aluminium cladding, adjacent to the weld bead, was not affected, a point which has since been investigated further.

Points (iv) and (v) are undoubtedly of importance if gas welding is employed because of the considerable spread of heat with this process. With the argon-shielded arc processes (tungsten-arc and self-adjusting arc), however, the heat source is intense and localized, hence welding is rapid and the heat-affected zone is considerably reduced. It was deduced from hardness surveys that the heat-affected zone, far from being reduced to the annealed strength, did in fact more nearly approach the solution heat-treated unaged condition. There also appeared to be adequate strength, if not ductility, in the over-heated zone adjacent to the weld bead, since after re-heat treatment joint efficiencies of almost 95% could be obtained. Since gas welding could be avoided it did not appear that points (iv) and (v) posed insuperable difficulties.

Even with tungsten-arc welding, however, the as-welded joints were weak and lacking in ductility. Clearly new filler materials were required which would give stronger weld metal. Further consideration suggested that the lack of ductility was partly a result of the weakness of the weld metal. Elongation measurements are normally made on a gauge length of 2 in., but the weld metal itself may not occupy more than one-quarter to one-eighth of this length. If the weld metal is weak and surrounded by much stronger parent metal, a condition may exist where the ultimate strength of the weld metal is less than the proof stress of the surrounding parent metal. Cast weld metal then has to provide the majority of the elongation. In the case just quoted, weld metal with an adequate ductility of, for example, 8% would probably give about 1% elongation in the joint. Measurements on small gauge lengths within the weld metal have actually confirmed that weld metal is not as lacking in ductility as elongation measurements on the joints appear to indicate. It was thought that if the weld metal was stronger, more yielding would occur in the plate and the overall ductility might be better. It appeared, therefore, that a solution to point (ii) would also go a long way to solving point (iii). Weld cracking with the high-strength aluminium alloys (point (i)) is undoubtedly serious, although it is considerably less with tungsten-arc welding than gas welding. Since the tendency to cracking depends mainly on composition the correct choice of filler material is clearly important. There was evidence to suggest that the cracking experienced with the Al–Cu–Mg alloys was little worse than that with the Al–Mg–Si alloys and it was considered that there was a reasonable chance of being able to develop a suitably crack-resistant filler alloy, at least for the Al–Cu–Mg alloys.

The above considerations pointed clearly to the development of new filler materials as the first step in the programme.

FILLER METAL REQUIREMENTS

To assist in the selection of alloys for filler materials, it was thought desirable to set standards of acceptance at an early stage. It was decided to aim for a weld-metal strength in the 'as-welded' condition of 25 tons/in² U.T.S. and in the fully re-heat treated condition of 30–35 tons/in² U.T.S. This may appear to be an unduly exacting requirement for a cast material, but it was known that weld metal, especially that deposited by the argon-shielded processes, was exceedingly good-quality cast material, exhibiting strength and ductility properties similar to the wrought material.

It was rather more difficult to set a standard for crack resistance, chiefly because there were no very precise methods for examining and reporting crack resistance in welds. The solution to this difficulty appeared to be the comparison of filler alloys under investigation with the known behaviour of the Al–Mg–Si alloy H.10, when welded with conventional filler materials. It was known that severe cracking was experienced with H.10, welded with parent-metal filler, but that if an Al–5% Si filler alloy was used only very occasional cracking was encountered. At first it was intended to design a restrained weld test jig which would crack H.10 welded with parent metal, but not crack H.10 welded with Al–5% Si filler. This was actually attempted but subsequently a more simple and reproducible test was developed which has proved invaluable in the present work.[6] From the behaviour of alloy H.10 in this test it was judged that filler alloys giving a crack length of less than 0·5 in. would be satisfactory.

A further requirement for the filler metal was that, to keep the corrosion risk to a minimum, it should not differ too much from the parent-metal composition.

FILLER METAL COMPOSITIONS

Several alloy systems appeared at first sight to offer reasonable scope for investigation, but the field had to be reduced because the combinations of various elements and percentages of elements were too numerous.

Three factors enabled the number of alloys for investigation to be limited. First, the best strength results in the preliminary tests had been obtained with

Table I
CRACKING TESTS WITH SPECIAL ALLOY FILLERS
D.T.D. 646 PARENT METAL

No.	*Filler alloy*	*Nominal filler composition, %*					*No. of tests*	*Cracking*			
								Centre		*Edge*	
		Cu	*Mg*	*Si*	*Zn*	*Ti*		*Length*	*No.*	*Length*	*No.*
19	1	6	—	—	—	0·10	6	0·18	2	1·04	6
20	3*A*	6	1·0	1·5	—	0·10	6	0·30	1	0·58	6
21	3*B*	6	1·0	1·0	—	0·10	6	0·35	3	0·76	6
22	4*A*	6	1·5	2·0	—	0·10	6	0·13	2	0·65	6
23	4*B*	6	1·5	1·0	—	0·10	6	0·40	3	0·52	6
24	578	4·4	0·5	0·8	—	0·15	6	0·30	2	1·07	6
25	580	4·4	0·5	0·8	—	0·15*	6	0·20	1	1·28	6
26	5% Si	—	—	5·0	—	—	6	0·36	6	0·65	6
27	10% Si	—	—	10·0	—	—	6	0·26	6	0·33	5
28	Cusil 410	4	—	10·0	—	—	6	0·21	5	0·70	4
29	Cusil 420	4	—	20·0	—	—	5	0·25	3	0·30	2
30	Cusil 83	8	0·35	3·0	—	0·25	7	0·17	5	0·44	5
31	Cusil 73 (*a*)	7	0·5	3·0	—	0·25	6	0·15	6	0·45	6
32	Cusil 73 (*b*)	7	1·0	3·0	—	0·25	5	0·20	5	0·50	4
33	Cusil 73 (*c*)	7	1·5	3·0	—	0·25	6	0·20	6	0·46	4
34	Cusil 72	7	0·2	2·0	—	0·50	7	0·20	6	0·70	7
35	Cusil 72	7	0·4	2·0	—	0·50	6	0·20	5	0·80	6
36	Cusil 72	7	0·8	2·0	—	0·50	5	0·18	3	0·85	5
37	Cusil 72	7	1·0	2·0	—	0·50	5	0·20	3	0·87	5

* Also contains 0·05% boron.
Dilution approx. 70%. Cracking results expressed in inches; either centre or edge cracks. Number of tests exhibiting each type of cracking is given in second column.
Parent metal analysis: Cu 4·15%, Mg 0·72%; Si 0·78%.

parent-metal filler alloys. Secondly, it was known that the rapid cooling during welding leaves the weld bead in a virtually solution heat-treated condition, so that it is susceptible to ageing. Thirdly, it had been shown that dilution effects were extremely important with the tungsten-arc welding of aluminium alloys.[7] (Dilution is the amount by which the filler material is diluted with melted parent metal in the making of the weld. Usually expressed as a percentage, dilution depends to a large extent on edge preparation, being a maximum with the square close butt joint. The weld bead composition has been shown to be substantially uniform so that it can be calculated from the weld bead geometry and the known analysis of parent metal and filler metal. It is the weld bead composition which is the important consideration).

For the first two reasons, just mentioned, it was decided to limit the investigation to the parent-metal alloy systems and to determine in particular the effect of increasing the various alloy and grain refining additions. It was known that because of the effect of dilution there would be little point in making only minor changes in filler alloy composition. At a later stage in the investigation, when certain trends had been established, it was possible to apply the knowledge of dilution effects to fix the limits of various additions.

TESTS FOR FILLER METAL

Filler alloys were assessed in three ways:

(*a*) By determining the crack sensitivity with the 'fishbone' test[6]

(*b*) By tensile testing welded butt joints in 12 s.w.g., $\frac{1}{4}$-, and $\frac{1}{2}$-in. thick plate

(*c*) By ageing tests on the 'as-deposited' weld metal, using bead-on-plate tests.

The ageing tests were carried out in the later stages of the programme and proved to be a most useful way of assessing weld metal properties. In transverse tensile testing of welds failure of the test piece in the heat-affected zone or in the overheated zone at the edge of the weld bead often masked the effect of changes in composition on weld metal properties. Failure in these areas is of course a separate issue influenced only indirectly by the weld bead and filler metal compositions. The ageing characteristics of the weld bead were studied by hardness surveys.

PREPARATION OF FILLER MATERIAL

Since a large number of different alloys would be be required in small quantities special consideration was given to the production of material in a suitable form for welding. Two methods were used, the first producing drawn rod, the second cast rod.

For most of the early work the alloys were cast to 1-in. diameter × 4 in. long ingots which were extruded in a specially designed inverse extrusion press fitted to a 50-ton tensile testing machine. Pressures up to 45 tons were used to produce $\frac{5}{16}$-in. dia. rod, which was then drawn to wire of a suitable gauge for the various tests.

Where the filler wire was required in only short lengths, such as for cracking tests, a casting technique was employed. The alloys were cast to 12 in. × 4 in. × $\frac{1}{8}$ in. thick sheet. Strips $\frac{1}{8}$ in. wide were cut from this sheet and the edges rounded by grinding to produce rod of $\frac{3}{32}$-in. diameter.

FILLER ALLOYS FOR WELDING Al–Cu–Mg ALLOYS

Investigation of Weld Cracking

As mentioned previously the aim was to discover whether filler alloys could be found in the Al–Cu–Mg

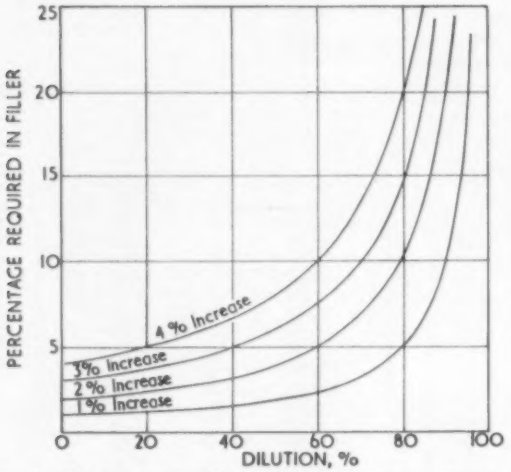

Fig. 1—Additions required to filler alloy to produce a given increase in weld metal composition at any dilution

system which would give both adequate strength and cracking resistance. Details of the crack-testing technique are not included as they have been given elsewhere.[6]

In the first series of tests (19–29 Table I) alloys based on 6% Cu with variations in silicon and magnesium were compared with parent-metal fillers and readily available fillers such as Al–5% Si, and Al–4% Cu–10% Si. An Al–4% Cu–20% Si alloy was also available in cast form from experiments on high dilution and a test with this filler material was also included in the series, as the alloy which would probably give maximum resistance to cracking.

The results given in Table I, (19–29), show that although edge-cracking (in the junction between weld metal and parent metal) is characteristic of the parent metal, it is still controllable by the filler metal. In this respect additions of silicon are particularly beneficial as was expected. Although the best results were obtained with the alloy containing 20% silicon a significant improvement was effected by additions of up to 2% silicon to the Al–6% Cu alloy. Filler No. 22 has the lowest combination of centre- and edge-cracking. The most severe cracking occurred with the parent-metal filler alloys 578 and 580 (Nos. 24 and 25) and it is surprising to note that more cracking was observed with the filler alloy containing boron although the grain refinement was more effective. Similar observations were made with the Al–Zn–Mg alloys to be discussed in a further report.

Large additions of silicon to a filler material were thought to be undesirable from the strength point of view and attempts were then made to reduce the silicon content to a minimum and still obtain the required resistance to cracking. It was at this stage in the investigation that the ageing characteristics of the weld metal were investigated and greater use was made of dilution data. A slight digression is necessary here to explain this in greater detail.

A Suitable Compromise

The work on ageing, reported in document *LM*8/25/58* was carried out to find ways of exploiting the fact that immediately after welding the weld bead is virtually in a solution heat-treated condition, and is susceptible to ageing. This aspect is discussed in a following section. The high-silicon alloys were difficult to age and it was necessary to strike a compromise between susceptibility to ageing and resistance to cracking.

An even more important compromise which had to be made was not fully appreciated until the completion of the first series of tests. This concerned the dilution which occurs in welding. Except where parent metal or similar filler metal is used the weld bead composition can vary considerably according to the edge preparation and the corresponding dilution. The final choice of filler alloy must therefore be a compromise giving satisfactory results over a certain range of dilution. Once the limits of dilution, between which the filler rod must be satisfactory, have been chosen, it should be possible, when knowing the trend of properties with composition, to select a filler composition which will give the desired result.

Choice of Filler Alloy according to Dilution

In most types of weld the dilution is between 25 and 70%. Only with square close butt joints in thick plate is the dilution much greater than 70%. There is therefore a strong case for specifying that with each type of parent alloy there should be at the most two filler alloys; one suitable over a range of dilution from 25–70% and a second for the special application of over 70% dilution. It would be difficult and possibly dangerous in practice to specify a number of filler rods each of which could be used over only a limited range of dilution. Slightly inferior results in certain instances must be accepted for greater simplicity in practical welding.

There is an additional reason why joints giving high dilution should be treated separately. The amount of any element in the filler alloy required to produce a given weld metal composition increases as the dilution increases and approaches infinity as the dilution approaches 100%. This is shown in Fig. 1, which also illustrates the difficulty of increasing the amount of any element in the weld bead by 3% more than the parent material contains. To add more than 3% to the weld bead it might be necessary to have total alloy additions to the filler metal in excess of 10% which would cause difficulties in the production of the wire.

The requirement that a filler alloy should be suitable for use between 25 and 70% dilution is considered a practical necessity and the subsequent discussion is devoted to the selection of filler alloys of this type.

Selecting the Silicon Content

Practical experience and laboratory tests show that Al–Cu alloys are made less susceptible to cracking by the addition of silicon. The cracking diagram for Al–Cu–Si alloys determined by Pumphrey *et al.*[4] suggests that there is little advantage to be gained in the region of 5% Cu by adding more than 2% silicon. This is convenient, since above this amount, weld properties suffer and the weld is noticeably sluggish in ageing. The filler rod should, therefore, contain

*To be published shortly in the *Journal*.

enough silicon to produce weld metal with about 2% silicon between the limits of dilution 25–70% and over the range of parent-plate composition normally encountered. The specifications for duralumin type alloys allow up to 1·5% silicon, but in practice it is unlikely that the silicon would be outside the limits 0·5–1·20%. To fulfil this requirement a filler alloy would have to contain about 3·5% silicon, as shown in Table III. This would give a maximum silicon content in the weld bead of 2·9% with 25% dilution in a parent metal containing 1·2% silicon. The minimum silicon would be 1·4% at 70% dilution in parent metal with 0·5% silicon.

Selecting the Copper Content

For both strength and resistance to cracking, it appeared desirable to keep the copper content of the weld bead on the upper limit, or even slightly above that permitted in the H.14 and H.15 alloy specifications.

The specification limits for the copper content of the parent plate are 3·5 to 4·8%, but in practice most duralumin-type alloys probably contain between 3·9 and 4·5% Cu. In selecting the copper content of the filler rod it should be noted that the alloy content is high, therefore any appreciable increase in weld metal copper content with 70% dilution will necessitate very large additions of copper to the filler rod, e.g. to produce Al–6% Cu weld metal from Al–4% Cu plate would require at 70% dilution an Al–11% Cu filler rod. This filler alloy with 11% Cu, when used at 25% dilution would produce weld metal with 9% Cu. Thus, not only is a filler alloy required which has an inconveniently high copper content for working into wire but the weld metal deposited at 25% dilution contains a dangerously high copper content.

A filler alloy containing 8% Cu appeared to be a good compromise giving in the weld a minimum of about 5·1% and a maximum of about 7·1% Cu depending on the dilution as shown in Table III.

Table II

ANALYSIS OF Al–Cu–Mg–Si FILLER ALLOYS

No.	*Filler alloy*	*Filler alloy composition, %*				
		Zn	*Mg*	*Cu*	*Si*	*Ti*
19	1	—	—	6·22	0·1	0·075
20	3*A*	—	1·02	5·92	1·45	0·10
21	3*B*	—	1·0	6·12	0·97	0·085
22	4*A*	—	1·46	5·98	1·95	0·070
23	4*B*	—	1·50	5·98	1·01	0·065
24	578	—	0·58	4·37	0·81	0·014
25	580	—	0·53	4·37	0·76	0·015*
26	5% Si	—	—	—	5·1	—
27	10% Si	—	—	—	10·3	—
28	Cusil 410	—	—	3·92	11·4	—
29	Cusil 420	—	—	3·04	20·5	—
30	Cusil 83	—	0·36	8·12	3·2	0·33
31	Cusil 73(*a*)	—	0·46	7·35	3·52	0·23
32	Cusil 73(*b*)	—	0·97	7·20	3·00	0·23
33	Cusil 73(*c*)	—	1·55	7·10	2·95	0·23
34	Cusil 72	—	0·20	6·93	1·92	0·76
35	Cusil 72	—	0·43	6·96	1·97	0·76
36	Cusil 72	—	0·82	6·62	2·29	0·64
37	Cusil 72	—	1·07	6·72	2·30	0·62

* Contains also 0·05% boron.

Table III

INFLUENCE OF Al–Cu–Mg–Si PARENT MATERIAL COMPOSITION AND DILUTION ON WELD COMPOSITION WHEN USING A FILLER ALLOY CONTAINING 8·0% Cu and 3·5% Si

Element	*Spec. limits of parent metal, %*	*Practical limits of parent metal, %*	*Dilution, %*	*Weld composition, %*
Copper	3·5–4·8	3·9–4·5	70 25 70 25	5·1 7·0 5·5 7·1
Silicon	Up to 1·5	0·5–1·2	70 25 70 25	1·4 2·7 1·9 2·9

Further Tests on Filler Alloys

With the above considerations in mind a further series of filler alloys was prepared for crack testing. On this occasion all the alloys were prepared by the cast-plate technique described previously. The 8% Cu, 3% Si alloy referred to above, Cusil 83 (No. 30), proved to be the most crack-resistant filler alloy with the exception of the alloy containing 20% silicon. There was, in fact, little to choose between the alloys for crack resistance and both satisfied the requirement of giving less than 0·5 in. cracking which had been established on the basis of the behaviour of the Al–Mg–Si alloy H.10.

Since it was found from ageing tests that the magnesium content of the Cusil 83 alloy (No. 30) was too low to give any marked ageing, a series of tests was made with three different magnesium contents, 0·5, 1·0, and 1·5%. In case the increase in magnesium content affected the cracking resistance adversely, cracking tests as well as ageing tests were carried out on these fillers. It is evident from tests 31, 32, and 33 that magnesium content has no effect on cracking.

When first used, the Cusil 83 filler was thought to be too fluid, although for light-gauge intricate work it was obviously very satisfactory. It appeared likely that a reduction in alloy content might help to alleviate this condition. A reduction in copper content from 8% to 7%, as in the tests 31–33, had apparently no noticeable effect on cracking and it was next decided to reduce the silicon content from 3% to 2%. To correct any tendency toward an increase in cracking resulting from this reduction in silicon content and to render grain refinement more effective, the titanium content was increased to 0·5%. When allowance was made for dilution it was expected that the titanium content of the weld bead would be about 0·25%. As a further check on the effect of magnesium on cracking at the lower silicon content, four levels of magnesium content were investigated. The results of tests with these alloys (34–37), showed that there was a marked increase in edge cracking to well above the 0·5-in. limit and it was concluded that the silicon content could not be reduced safely below 3%.

From the cracking point of view it appeared, therefore, that an alloy with 7% Cu and 3% silicon would be satisfactory for welding the Al–Cu–Mg alloys. Further practical experience with the alloy indicated

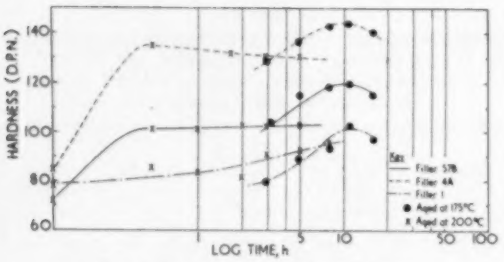

Fig. 2—Ageing behaviour of weld beads

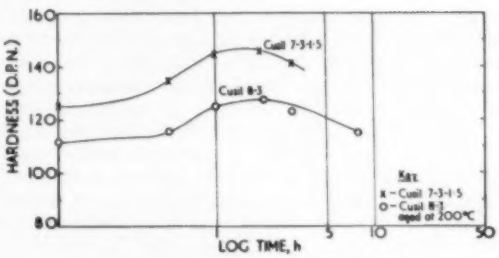

Fig. 4—Ageing behaviour of weld beads

that the high fluidity which resulted in reduced penetration in manual welding could be overcome by a suitable technique involving a slightly modified edge preparation. Since the magnesium content appeared to have no effect on cracking there was complete freedom to fix the limits for this element to suit ageing and mechanical property requirements.

Ageing Tests on Al–Cu–Mg Weld Beads

In arc welding, weld metal is cooled at a very high rate from the molten weld pool temperature, so that constituents are to a large extent retained in solution. It was known that, because of this, the properties of welds in the heat-treatable alloys could be improved by a simple ageing treatment. Even when filler metal is used, which in itself is not susceptible to heat treatment, sufficient alloying elements are usually picked up from the parent metal to make the weld bead susceptible to heat treatment or ageing.

To exploit this property and obtain joints of maximum strength, information was required on the effect of composition on the ageing behaviour of weld metal deposited using a variety of filler metals. The examination was carried out by the hardness testing of bead-on-plate samples since the testing of actual welds would have been too costly in time and material. This technique also eliminated the masking effect of weld defects which occurred with tensile testing.

It was found that suitably aged 'as-deposited' weld metal from several of the special fillers employed in previous work could reach hardnesses approximating that of the unaffected parent metal. This was thought to be of considerable importance because although large welded structures cannot generally be completely re-heat treated, there are many applications where artificial ageing after welding is practicable. The possibility of welding material in the solution heat-treated condition and ageing the whole assembly appeared promising.

Al–Cu–Mg weld beads respond only a limited extent to natural ageing. After 24 weeks the hardness is much below the hardness of the fully heat-treated D.T.D. 646 parent material and it appears that in practice, it would be necessary to resort to artificial ageing to obtain the maximum weld bead strength.

A summary of the ageing behaviour of weld beads made with a number of the special filler alloys is given in Tables IV and V, and in Figs. 2–5. The parent-metal filler 578, No. 24, gave a maximum hardness of 123 after ageing at 160°C and a hardness of 102 after ageing at 200°C. These figures were exceeded by several of the special alloys notably those with high magnesium. From the hardness point of view it appears that a filler alloy containing not more than 1% magnesium and possibly 0·6–0·8% would give a satisfactory over-matching of the weld bead compared with the heat-affected zone. The alloys Cusil 83 and Cusil 73 do not show a very marked increase in hardness as a result of artificial ageing, but they do give a higher initial hardness which could be an advantage where artificial ageing cannot be carried out. Hardness is not the only criterion, of course, and other properties are also important. Having arrived at what appeared to be a suitable composition from cracking tests and ageing tests it was now necessary to carry out an extensive programme of mechanical and fatigue testing, together with practical trials and corrosion tests.

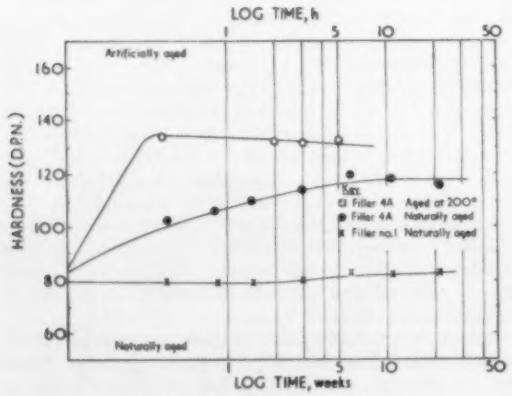

Fig. 3—Ageing behaviour of weld beads

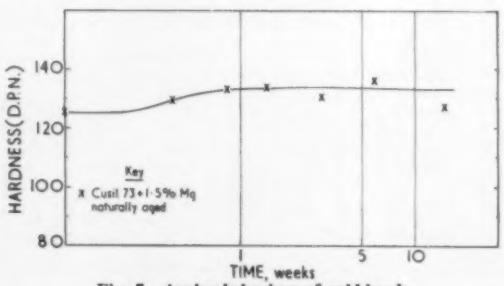

Fig. 5—Ageing behaviour of weld beads

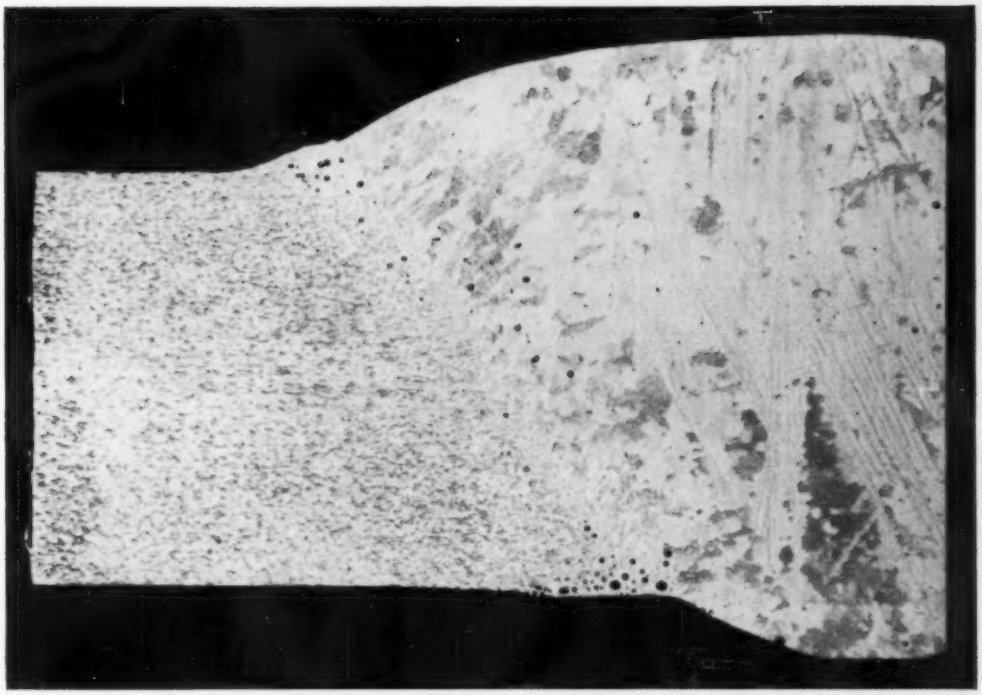

Fig. 6—Fusion-zone porosity in ¼-in. thick argon-arc weld in Al–Cu–Mg alloy H.15 using 'Cusil 83' filler

Tensile Tests on Welds in Al–Cu–Mg Alloy

In welds in high-strength aluminium alloys there is a general tendency for the ultimate tensile strength to drop as the thickness welded increases. This may be due to several causes.

(i) Thick material is welded more slowly than thin, so that the strength of the heat-affected zone is reduced

(ii) For the same reason the weld bead is cooled less rapidly and there is a reduced tendency for constituents to be retained in solution so that the weld bead is less susceptible to ageing

(iii) Again, for the same reason as in (i), the overheated zone is more extensive

(iv) The structure of the weld bead is more coarse and defects, particularly micro-porosity, can assume a more serious form.

(v) Dilution becomes less, so that unless special filler materials are used the weld bead may be depleted in constituents promoting age-hardening.

Table VII shows the effect of plate thickness on the strength of argon-arc welds made with 'standard' filler materials. The general trend towards lower properties with the ½-in. thick material is evident. The Al–Cu and Al–Mg filler alloys give much reduced properties in the ½-in. thickness, particularly after

Table IV

ARTIFICIAL AGEING OF Al–Cu–Mg–Si WELD BEADS

No.	*Filler alloy*	*Ageing time to reach maximum hardness at various temperatures, h*				*Maximum hardness attained at various ageing temperatures*			
		200°	*190°*	*175°*	*160°*	*200°*	*190°*	*175°*	*160°*
19	1	8	11	11	16	94	95	101	100
20	3*A*	½	5	8	22	126	136	139	140
21	3*B*	1	10	11	22	122	126	129	135
22	4*A*	½	4	10	22	135	136	144	148
23	4*B*	½	6	11	22	131	138	138	141
24	578	½	11	11	17	102	115	120	123
26	5% Si	½	4	16	16	81	89	96	97
27	10% Si	½	3	11	16	95	102	120	113
29	Cusil 420	—	—	—	—	140	140	140	140
30	Cusil 83	2	5	16	22	127	124	125	129
33	Cusil 73	2	—	—	—	146	—	—	—

Table V

NATURAL AGEING OF Al–Cu–Mg–Si WELD BEADS

No.	*Filler alloy*	*Hardness after 12 weeks*	*Hardness after 24 weeks*	*Maximum hardness after ageing at 200°C*
19	1	82	82	94
20	3*A*	112	112	126
21	3*B*	111	112	122
22	4*A*	117	117	135
23	4*B*	115	115	131
24	578	99	100	102
26	5% Si	71	71	81
27	10% Si	81	81	95
29	Cusil 420	140	140	140
30	Cusil 83	117	117	127
33	Cusil 73	133	133	146

re-heat treatment of the weld, so that point (v) above clearly has a considerable influence on mechanical properties. The special filler materials were designed to overcome this defect. In the $\frac{1}{8}$-in. thick tests the prolonged heating during welding has caused severe softening in the heat-affected zone and in three cases failure took place in the heat-affected zone.

Welds in $\frac{1}{4}$-in. Material

A series of argon-arc welds was made in $\frac{1}{4}$-in. D.T.D. 646 using the special filler alloys, Nos. 19–25, Table I.

In the as-welded condition, the joint efficiencies of 63–72% were higher than had been obtained previously with the simple binary fillers Al–5% Cu, Al–5% Si, and Al–6% Mg.

The results were all of the same order and it is probable, therefore, that the improvement was mainly a result, as shown in Table VIII, of a slightly higher copper content and better grain refinement. Alloys 3*B* and 4*B* gave slightly higher results than alloys 3*A* and 4*A*, which have higher silicon but are otherwise the same. Dilution effects have tended to level out composition differences and it is clearly impossible to relate properties to weld composition satisfactorily.

After heat treatment all the welds gave strengths of approximately 30 tons/in² with failure in the plate material approximately $\frac{1}{2}$ in. from the weld centre. Parent-plate failures at only about 90% joint efficiency suggest that the heat treatment after welding was imperfect.

The location of fracture in the as-welded condition was usually at the edge of the weld when the strength was over 21·0 tons/in². This type of failure was caused by the development of spherical porosity in the partly melted parent plate.

Weld in $\frac{1}{2}$-in. Material

Using the same filler alloys, tests were made on

Table VI

Al–Cu–Mg–Si WELD BEAD COMPOSITIONS IN AGEING TESTS

No.	*Dilution, %*	*Weld composition, %*			
		Zn	*Mg*	*Cu*	*Si*
19	27	—	0·2	5·6	0·2
20	28	—	0·9	5·4	1·3
21	40	—	0·9	5·3	0·9
22	36	—	1·2	5·3	1·5
23	45	—	1·2	5·2	0·9
24	34	—	0·6	4·3	0·8
26	46	—	0·3	2·0	3·1
27	30	—	0·2	1·3	7·2
29	30	—	0·2	3·4	14·5
30	23	—	0·5	7·2	2·6
33	28	—	1·30	6·3	2·3

Table VII

EFFECT OF MATERIAL THICKNESS ON MECHANICAL PROPERTIES OF ARGON-ARC WELDS IN HIGH-STRENGTH ALLOYS

Type of parent metal	*Test condition*	*Property*	*12 s.w.g.*			*$\frac{1}{4}$-in. thick*			*$\frac{1}{2}$-in. thick*		
			Filler alloy			*Filler alloy*			*Filler alloy*		
			PM	*Mg*	*Cu*	*PM*	*Mg*	*Cu*	*PM*	*Mg*	*Cu*
Al–Cu–Mg (D.T.D. 646)	As-welded	U.T.S., tons/in²	21·4	18·8	18·5	19·5	18·3	16·1	16·8	17·7	15·9
		Elong. on 2 in.	3	3	3	4	3	3·5	5	5	5
		Location fracture	WM	WM	WM	WM	WM	WM	PM	PM	WM
Al–Cu–Mg (D.T.D. 646)	Re-heat treated	U.T.S., tons/in²	30·0	24·1	29·4	29·8	28·1	27·7	24·2	24·2	19·1
		Elong. on 2 in.	2	3	3	2·5	2·5	1·5	2	3	1
		Location fracture	WM	WM	WM	WM	WM	WM	WM	WM	WM
Al–Zn–Mg (D.T.D. 683)	As-welded	U.T.S., tons/in²	22·9	22·1	—	20·2	20·0	—	20·8	19·7	—
		Elong. on 2 in.	1·5	2·0	—	2·0	2·0	—	5·0	5·0	—
		Location fracture	WM	WM	—	WM	WM	—	PM	WM	—
Al–Zn–Mg (D.T.D. 683)	Re-heat treated	U.T.S., tons/in²	36·3	29·9	—	35·7	32·1	—	34·9	23·8	—
		Elong. on 2 in.	4	1·5	—	3	2	—	4	4	—
		Location fracture	WM	WM	—	WM	WM	—	WM	WM	—

Key to filler alloy *PM* = Parent metal
WM = Weld Metal
Mg = Al–6% Mg
Cu = Al–5% Cu
All figures are average of 3–5 tests.

Dilution in 12 s.w.g. sheet 70–75% single-pass
" " $\frac{1}{4}$-in. plate 40–50% single-pass
" " $\frac{1}{2}$-in. plate 45% double-pass
All welds made manually—machined before testing.

Table VIII

STRENGTH OF WELDS IN 12 S.W.G. D.T.D. 646 SHEET

Filler alloy	*Dilution, %*	*As-welded properties*					*Heat-treated properties*				
		U.T.S., tons/in²		*Fracture**			*U.T.S., tons/in²*		*Fracture*		
		Range	*Average*	*Weld*	*E. of W.*	*Plate*	*Range*	*Average*	*Weld*	*E. of W.*	*Plate*
1	72	20·8–22·1	21·2	4	—	—	29·5–32·9	30·6	—	—	5
3*A*	76	19·8–21·4	20·7	2	2	—	—	33·8	—	—	1
3*B*	77	22·0–23·6	22·3	—	5	—	28·4–32·6	30·5	—	—	5
4*A*	67	20·8–21·7	21·0	1	5	—	28·4–30·0	29·3	—	—	4
4*B*	72	20·8–22·2	21·5	—	4	—	27·7–30·7	29·7	—	—	5
580	69	20·4–22·8	21·4	—	6	—	26·9–32·1	30·0	—	—	6

*Figures indicate number of samples and location of fracture.
Parent-plate properties: D.T.D. 646—32·5 tons/in² U.T.S., 11% elong. on 2-in. gauge length.
Elongation values were 2–5% for all the welds quoted.
Welding technique: argon, square close butt joint.
Parent-metal composition: Cu 4·39%, Mg 0·46%, Si 0·85%, Mn 0·85%, Fe 0·37%.

welds in ¼-in. plate. These welds were tested in three conditions: as-welded and naturally aged; as-welded and artificially aged; and completely re-heat treated.

From Table IX, in which the results are summarized, it will be seen that in the as-welded condition the ultimate tensile strengths are considerably lower than those obtained with the ⅛ in. thick materials, e.g. 17 to 19 tons/in² compared with 21 to 22 tons/in² with the thinner material. The best results were obtained with the filler alloy No. 1 containing 6% copper. Welds made with the other filler alloys may have had a lower strength because of the rather higher silicon content which tended to inhibit natural ageing. There was little difference in the ultimate tensile strength obtained with the filler alloys 3*A*, 3*B*, 4*A*, 4*B* and the parent-metal filler 578. In the as-welded condition the location of fracture was invariably in the centre of the weld metal.

Welds made with the filler alloys 3*A*, 3*B*, 4*A*, and 4*B* showed a useful increase in ultimate tensile strength after artificial ageing, e.g. for filler 3*A* from 17·4 to 19·4 tons/in². As a result of the ageing treatment the location of fracture was moved in many cases to the heat-affected zone alongside the weld, but it was clear from hardness surveys that the parent material had been severely over-aged. The prolonged ageing time was adopted to get the maximum strength from the weld metal irrespective of parent-plate condition.

After full re-heat treatment, strengths obtained ranged from 26·75 to 28·6 tons/in², for the special alloy fillers and about 29 tons/in² for the parent-metal filler, with the location of fracture in the centre of the weld metal.

Tests with 3% Silicon Crack-resistant Alloys

Although the tests just described left many questions unanswered, no attempt was made to repeat the series since it had been decided by this time that the filler materials which had been used were not sufficiently resistant to cracking. A further series of tests was therefore carried out on the higher silicon alloys (Nos. 30–37, Table I). Tests in 12 s.w.g. and ¼-in. D.T.D. 646 with the filler alloy Cusil 83, which appeared from dilution considerations to be suitable as regards cracking resistance, are given in Table X.

It was hoped that by welding parent metal in the solution heat-treated condition and then ageing the welded joint, the maximum weld bead strength would be obtained with the least over-ageing of the parent metal. An ageing treatment of 10 h at 175°C which was known to give the highest hardness in the weld

Table IX

AVERAGE MECHANICAL PROPERTIES OF WELDS IN ¼-IN. D.T.D. 646 PLATE

Filler alloy	*Ultimate tensile strength*			*Location of fracture**		
	As-welded	*Welded and aged†*	*Re-heat treated*	*As-welded*	*Welded and aged*	*Re-heat treated*
1	19·0	20·0	27·1	CW(4)	CW(3) AZ(1)	CW(2)
3*A*	17·8	19·4	27·1	CW(4)	CW(1) AZ(1)	CW(3)
3*B*	17·6	18·7	26·8	CW(3)	AZ(2)	CW(2)
4*A*	17·5	18·5	28·4	CW(4)	EW(2) CW(1)	CW(1) EW(1)
4*B*	17·3	18·9	28·6	CW(2)	CW(2)	CW(1) EW(1)
578	17·3	19·2	29·5	CW(4)	AZ(2) CW(1)	CW(2)
580	16·9	19·5	22·7	EW(3) CW(1)	EW(2) AZ(1)	EW(2)

* Figures in brackets indicate the number of tests. † Aged for 7 h at 185°C.
CW = centre of weld; EW = edge of weld; AZ = annealed zone.
Parent-metal analysis: Cu 4·15%, Mg 0·72%, Si 0·78%, Mn 0·72%, Fe 0·41%.
Dilution: 40–50%.
Welding technique: Argon-arc, single-pass, V-butt over backing bar.

Table X

STRENGTH OF ARGON-ARC WELDS IN D.T.D. 646 MADE WITH FILLER ALLOY CUSIL 83

Plate thickness	*Plate condition*	*Treatment after welding*	*Average U.T.S., tons/in²*	*Elong. on 2 in.*	*Location of fracture*			*No. of tests*
					Weld	*E. of W.*	*Plate*	
12 s.w.g.	S.T.	10 h at 175°C	19·9	2·0	1	7	—	8
	S.T. and F.H.T.	Re-heat treated	29·3	3·0	4	3	—	7
	F.H.T.	1 h at 200°C	19·8	2·0	—	7	—	7
¼ in.	S.T.	10 h at 175°C	19·9	6·5	—	—	7	7
	S.T.	Re-heat treated	28·8	3·0	—	4	—	4
	F.H.T.	1 h at 200°C	19·7	2·0	—	4	—	4

S.T. solution-treated 50 min at 508°C W.G. F.H.T. fully heat-treated, as above, then 2 h at 180°C.
Dilution: 12 s.w.g. sheet 75%; ¼-in. plate 45–50%.
Edge preparation, single-pass V-butt.
Filler alloy analysis: Cu 8·12%, Mg 0·36%, Si 3.20%, Mn 0·4%, Fe 0·2%.
Parent-metal analysis as in Tables VIII and IX.

bead, was selected. The treatment was successful with the welds in ¼-in. thick plate which failed in the heat-affected zone at about 20 tons U.T.S. giving a good elongation. In the 12 s.w.g. material there had been less over-ageing in the heat-affected zone and the weakest point was the edge of the weld.

Welds made in fully heat-treated plate were given a shorter ageing treatment of 1 h at 200°C. This ageing treatment did not raise the weld bead to maximum hardness and failure occurred at the edge of the weld, this was the most common type of failure. Joints fully heat-treated after welding gave efficiencies of 90%.

Although the U.T.S. results were no higher than had been obtained previously, with parent-metal filler for example, it was the first time such strengths had been achieved with a crack resistant filler material. Since the ageing tests had shown that the magnesium content could be usefully increased, a series of tensile tests

Table XI

PROPERTIES OF WELDS IN D.T.D. 646 MADE WITH Al–Cu–3% Si FILLER ALLOYS

Test No.	*Filler alloy*	*Weld metal composition**			*Condition of plate*	*After-treatment*	*U.T.S., tons/in²*		*Elong. 2 in.*	H_v	*Fracture†*		
		Cu	*Si*	*Mg*			*Range*	*Average*			*Weld*	*E. of W.*	*Plate*
	12 s.w.g. plate						*12 s.w.g. plate*						
1	Cusil 83	5·6	1·8	0·7	F.H.T.	F.H.T.	20·0–25·0	22·5	2·5	1·03	1	5	—
2	Cusil 63	4·8	1·7	0·7			21·7–26·4	24·6	3·0	1·05	2	4	—
3	Cusil 73 (*a*)	5·1	1·6	0·6			23·6–25·0	24·3	2·0	1·02	2	4	—
4	Cusil 73 (*b*)	5·2	1·5	0·8			21·4–25·1	23·9	1·5	1·03	1	5	—
5	Cusil 73 (*c*)	4·9	1·3	0·9			20·7–26·0	24·0	2·0	1·03	—	6	—
6	Cusil 83	5·6	1·8	0·7	F.H.T.	N.A. 6 weeks	18·5–19·5	19·0	3·5	1·16	—	3	—
7	Cusil 63	4·8	1·7	0·7			18·8–23·9	19·2	3·0	1·08	—	2	—
8	Cusil 73 (*a*)	5·1	1·6	0·6			17·9–19·0	18·3	2·0	1·16	—	3	—
9	Cusil 73 (*b*)	5·2	1·5	0·8			17·0–20·3	18·4	2·5	1·16	—	3	—
10	Cusil 73 (*c*)	4·9	1·3	0·9			19·2–20·7	19·95	2·5	0·99	—	2	—
	¼-in. plate						*¼-in. plate*						
11	Cusil 83	5·7	1·8	0·7	F.H.T.	F.H.T.	23·7–25·8	24·6	2·0	0·87	1	4	—
12	Cusil 63	4·8	1·7	0·7			23·0–24·5	23·8	1·5	0·95	2	3	—
13	Cusil 73 (*a*)	5·3	1·7	0·6			26·1–26·9	26·6	2·0	1·07	—	3	—
14	Cusil 73 (*b*)	5·4	1·6	0·8			23·5–26·1	25·1	2·5	1·08	1	5	—
15	Cusil 73 (*c*)	5·5	1·8	1·0			24·0–25·6	25·0	3·0	1·17	—	8	—
16	Cusil 83	5·5	1·7	0·7	F.H.T.	N.A. 6 weeks	17·6–19·1	18·4	2·5	1·14	—	2	—
17	Cusil 63	4·9	1·7	0·7			17·4–18·8	18·1	2·0	1·09	—	4	—
18	Cusil 73 (*a*)	5·3	1·7	0·6			19·7–19·8	19·75	2·0	1·11	1	2	—
19	Cusil 73 (*b*)	5·3	1·6	0·8			18·8–19·0	18·85	1·5	1·1	3	2	—
20	Cusil 73 (*c*)	5·5	1·8	1·0			18·4–19·6	19·1	2·0	1·23	3	2	—
21	Cusil 63	4·9	1·7	0·7	S.T.	90 min 180°C	18·3–20·6	19·4	2·5	1·16	8	—	—
22	Cusil 73 (*a*)	5·3	1·7	0·6			17·3–21·0	19·7	2·5	1·2	2	4	—
23	Cusil 73 (*b*)	5·4	1·6	0·8			19·8–22·8	20·6	3·5	1·46	2	2	2
24	Cusil 73 (*c*)	5·1	1·5	1·0			19·1–20·8	20·1	4·0	1·5	2	2	2

* Estimated from dilution. † Figures indicate number of tests which failed as indicated.
F.H.T. = fully heat-treated. N.A. = naturally aged. S.T. = solution-treated.
Parent-metal analysis as in Tables XIII and IX

was then carried out on welds made with the higher magnesium fillers (tests 31–33). For comparison purposes welds with Cusil 83 filler and a similar alloy with 6% Cu, Cusil 63, were repeated. In these tests which are summarized in Table XI an attempt was made to obtain parent-metal failure after re-heat treating the joints by limiting the ageing treatment. It was hoped that the parent metal would then be slightly weaker than the weld metal so that failure would occur away from the weld. The treatment was not successful because the weld bead and over-heated zone failed to develop sufficient strength. Actually the results after full heat treatment are inferior by about 4 tons/in² compared with previous tests. It might be possible in welds with lower dilution to adjust the heat treatment to make the weld bead stronger than the parent metal without a serious reduction in overall properties, but each type of joint and application would require individual examination.

Table XI includes a column, H_u, for a factor which is the ratio of the lowest hardness in the weld bead to the lowest hardness in the heat-affected zone. When the factor is greater than 1·0 the hardness of the weld bead over-matches the parent plate. Along with other data on mechanical properties the H_u factor gives a useful indication of the uniformity of the joint, the condition of the weld bead, and the probable location of fracture. In the tests in ¼-in. plate aged for 90 min at 180°C (Table XI), the U.T.S. is approximately the same for welds with the four filler alloys but the H_u factor increases from 1·16 to 1·5 as the magnesium content is raised. At the same time the location of fracture tends to move towards the plate. It seems clear from these results and those in the previous table that the full strength of the weld metal is not being realized because of a weakness at the edge of the weld. This weakness appears to be due to fine porosity developed in the over-heated zone at the edge of the weld, and is possibly associated with the gas content of the parent material (*see* Fig. 6). Further investigation of this defect appears desirable now that the weld-edge failure has become a limiting factor to the strength of the joint. Although the 3% silicon alloys used for the tests in Table XI respond only slightly and very sluggishly to natural ageing, reasonable weld strengths were obtained in the naturally-aged tests because of the high initial hardness of the weld bead (*see* Figs. 4 and 5).

Because of the tendency for welds to fail at the edge, the artificial ageing of joints after welding had little effect on the U.T.S. of the joint. The artificial ageing treatment can be used, however, to move the fracture to the heat-affected zone by causing further softening in this area. With manual welding, therefore, the softening in the heat-affected zone is not necessarily a disadvantage. High-speed automatic welding in thin material may alter the situation, however, because the cooling rate and the time at high temperature will affect weld bead strength, the amount of softening in the heat-affected zone and to some extent the porosity in the over-heated zone. Which of these three factors would then limit the joint strength would depend on the particular welding conditions adopted.

Since most of the filler alloys used in the tests shown in Table XI were only available in cast form, the amount of welding which could be carried out was limited. There is clearly a need for further tests, preferably with one type of filler alloy, to investigate variations in welding conditions and after-treatment. The behaviour of the Cusil 73(*b*) alloy containing 1% Mg appears to be comparable with that of the Cusil 73(*c*) alloy containing 1·5% Mg, so that it would probably be satisfactory in future tests to limit the magnesium content of the alloy to 0·8–1·0%.

CONCLUSIONS

The difficulties in welding high-strength heat-treatable aluminium alloys have been examined. It was concluded that the most urgent problem requiring solution was the development of filler materials giving both crack resistant and high-strength weld metal. The two requirements are to some extent conflicting and it has been necessary to find a compromise.

A minimum standard of crack resistance was set at 0·5 in. of cracking in a newly developed cracking test. This standard was arrived at by comparison with the known behaviour of the Al–Mg–Si alloy H.10. An alloy giving a crack length of less than 0·5 in. in the test should behave at least as well as the Al–5% Si filler would with H.10 parent-metal alloy. For welding Al–Cu–Mg–Si alloys of the H.14 and H.15 types, a filler material containing 3% Si and about 7% Cu has been found to meet these requirements [patent applied for].

It was hoped to obtain weld metal with an ultimate tensile strength of 25 tons/in² in the as-welded condition. Although strengths of this order have not been obtained in transverse tensile tests, the 7% Cu, 3% Si alloy with about 1% Mg has given strengths of about 20 tons/in² in the as-welded condition. With this filler alloy, weld metal properties are maintained more easily as the thickness welded increases. Using a simple Al–Si filler alloy, weld metal properties are noticeably inferior in thicker material due to the decreased dilution.

Failure occurs generally at the edge of the weld in the over-heated zone and it is possible that if the weakness were overcome higher strengths could be obtained. Artificial ageing after welding will raise the weld bead hardness and with suitable treatment failure in the plate material with improved elongation may be obtained.

Over-ageing or softening in the heat-affected zone does not limit the weld joint strength at present.

Although more work must be done, particularly on the edge of weld failure and on mechanical properties, the results so far are sufficiently encouraging to suggest the application of welding to the Al–Cu–Mg alloys in carefully chosen applications.

Note: Work on the Al–Zn–Mg alloys has been carried out along similar lines to that just described. In terms of strength the results are more encouraging than for the Al–Cu–Mg alloys. A report is being prepared on the Al–Zn–Mg alloys which will form a continuation of the present document.

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Hydrogen in Mild-steel Weld Deposits

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SYNOPSIS

IMPROVED methods and equipment for hydrogen determination are described, and various sources of systematical errors are discussed. Evolution of CO is shown to be an important factor in the extraction at 700°C and at higher temperatures of samples containing a skin of oxide.

The concentration of hydrogen in single beads has been found to increase slightly from the starting point towards the end crater, with a pronounced increase in, and close to, the crater itself. Metallographic studies of gas bubble formation from beads sectioned and ground at low temperatures indicate a higher concentration of hydrogen in the centre than at the surface and fusion line. These gradients have been confirmed by micro-analytical measurements. Both the longitudinal and traverse gradients may be interpreted as a result of high-temperature diffusion.

Preliminary studies of micro-segregation of hydrogen indicate a local enrichment of hydrogen close to the austenite grain boundaries.

Introduction

EXPLORATORY measurements of hydrogen contents and internal micro-cracking have already been reported.[1] The present report deals with a more detailed investigation of the problems of sampling and measurement in hydrogen determinations. A separate progress report will be submitted on extended studies of microfissuring under various conditions of cooling and restraint.

Minor adjustments have been made in the original programme of work so as to collaborate in a joint programme recommended by the Hydrogen Sub-Committee (2A) of the I.I.W. In particular, several institutions have co-operated in investigating analytical procedures.

Much of the experimental material to be presented has been communicated to this Sub-Committee and a large part of the results of work on spatial distribution of hydrogen in single-bead weldments was presented at the public session of the I.I.W. annual meeting in Essen.

ANALYTICAL DETERMINATION OF HYDROGEN

The preliminary measurements reported previously for deposits from commercial rutile and basic electrodes were mainly intended to disclose whether there was a correlation between the gas contents of those deposits and their tendency towards microfissuring. The apparatus used at that time did not permit a chemical analysis of the extracted gases, and the total gas contents were reported with reservation that corrections might be required due to the presence of carbon monoxide or nitrogen. Subsequent work has shown that a substantial correction is sometimes needed. The conditions leading to contamination of the extracted gases are discussed later.

Methods and Apparatus

The specimens are shown in Fig. 1. The weldment was quenched in cold water immediately after completion of welding, and subsequently in alcohol–solid carbon dioxide. Further cooling in liquid nitrogen was used for storing when a series of ten samples was to be analysed for assessment of the hydrogen distribution along the axis of a bead. This precaution was also taken with the specimen shown in Fig. 1*c*, but not in routine determinations of the average hydrogen content, when the arrangement of Fig. 1*b* was used. This arrangement is the one adopted by the I.I.W.

Samples were obtained by breaking apart, at a very low temperature, the previously ground and weighed pieces. Each piece had been de-gassed for 72 h at 250°C before use so as to eliminate corrections for parent-plate hydrogen. Vacuum-extraction analysis at 650°C has shown an average hydrogen content after this treatment of about 0·013 ml of H_2 per 100 g, or about 1 micro-litre in each of the slices shown in Figs. 1*a* and *b*. The main part of the analytical work was done with a parent-plate material of the following nominal percentage composition:

C	*Mn*	*Si*	*S*	*P*	*N*
0·05	0·43	0·005	0·03	0·04	0·010

The amount of deposited metal is easily obtained with a fair degree of precision by weighing, whilst the total amount of fused metal is subject to much larger errors in the measurement of the fusion-zone contours.

2nd Progress Report on Project B 585, submitted to the Royal Norwegian Council for Industrial and Scientific Research, October 1957.

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The main parts of the analytical system are shown in Fig. 2. Samples may be charged into the apparatus, either directly into the hot extraction compartment through a mercury lift, or into the cold extraction chambers. In the direct-loading method, any foreign gas occluded with the sample was collected and measured in the McLeod system before heating. Small amounts of hydrogen in these gases were determined by means of a palladium osmosis tube maintained at 550°C. During the hot-extraction period measurements were made at 5-min intervals until the evolution of gas had practically ceased. In cases of abnormal behaviour to be described shortly, where no definite end-point is observed, extraction was discontinued after 2h. The moisture content was obtained by difference after circulation over P_2O_5, and the content of hydrogen was measured directly after diffusion through the palladium tube. The copper-oxide furnace, the ascarite–P_2O_5 absorption train, and the storage volumes were not used in routine analyses.

When the amounts of diffusible and residual hydrogen were to be measured separately, as was usually done, the samples were introduced with a magnet through mercury seals into the cold extraction chambers (item 12, Fig. 2). The specimens were left floating on mercury by lowering the level inside the chamber under suction. The outside limbs of the extraction chambers were vented to atmosphere at intervals to check the volumes of collected gas. When the rate of evolution had become negligibly small, the extracted gases were transferred to the main vacuum system for a final and more precise measurement in the McLeod gauge, and their hydrogen content was determined by diffusion through the palladium tube. Usually a value between 90 and 98% was obtained in this operation, the remainder being discarded as foreign gas entrapped with the sample. As a precaution against a possible catalytic oxidation of hydrogen by entrapped air, which might conceivably occur on the hot palladium wall, the open limb of the diffusion chamber was covered with a blanket of dry nitrogen while the specimen was being introduced. This precaution was also beneficial in preventing condensation of moisture on the chilled specimen during transfer from the cooling bath to the analytical apparatus.

When the sample had been de-gassed sufficiently for safe handling outside the apparatus, it was removed from the cold extraction chamber with a magnet. After weighing and light grinding to remove the oxide skin formed during welding, residual hydrogen was determined as for the direct-loading method.

Systematical Errors of Analysis

Systematical errors arising from weld-metal heterogeneity and from loss of hydrogen during the preparation of the sample will be considered in subsequent sections. The sources of error now to be discussed are the factors that may cause contamination of the hot-extracted gases.

The abnormal behaviour in hot extraction already mentioned is illustrated in Fig. 3. Curve *a* has been constructed from data obtained with an apparatus that was not equipped for chemical analysis of the extracted gases. The lack of a well-defined end-point and the parabolic course of the gas evolution indicate a process approaching its equilibrium state at a much slower rate than in the normal extraction of hydrogen. Processes of this kind occurring in the parent-plate material can easily affect the result, because the ratio of parent plate to weld deposit is quite high (between 9 and 10).

Curve *b* shows the result obtained in a control determination with the same materials; most of the hydrogen was removed in room-temperature extraction. After a period of five days under vacuum the specimens were ground lightly to remove any visible oxide skin formed during the welding. This precaution greatly reduced the rate of gas evolution in the subsequent hot extraction. It appears therefore, that the large amounts of gas indicated by curve *a* must be associated with the thin layer of oxide that is always present adjacent to the bead. The foreign gases derived from this layer might consist partly of absorbed moisture; however, even very strongly bound surface water could hardly account for the evolution of gas after more than 2 h at 650°C.

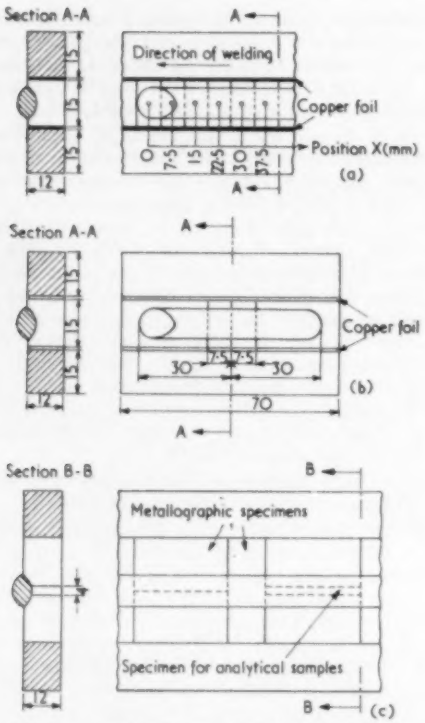
Further information of the composition of the foreign gas is presented in Fig. 4 and Table I.

Curve *a* shows the course of extraction at 650°C with a sample consisting of about 4 g of a basic electrode deposited on about 40 g of parent plate. The procedure was slightly different from the usual direct-loading method in that the sample had been stored in the vacuum apparatus for 21 h at room temperature before the hot extraction. A substantial fraction of the hydrogen measured after hot extraction must have been released during this first period. Accordingly, there was little hydrogen left for reaction with surface oxides, as is indicated by the rather small amount of water vapour in the extracted gas. Although it is not possible to infer which fraction of water vapour was present as such and which was formed by reduction of iron oxide, it may be stated that the former is not a significant source of error when the sample has been properly washed and dried. This conclusion has been confirmed repeatedly in later work.

The rate of gas evolution was not altered materially on prolonged extraction at 650°C. This is shown in curve *b* which, apart from an inevitable delay during the period of re-heating, forms a smooth continuation of curve *a*. An additional amount of only 4 micro-litres of hydrogen was recovered in one hour. The gas was mainly carbon monoxide, and was, of course, completely dry. The specimens were still tarnished after this treatment.

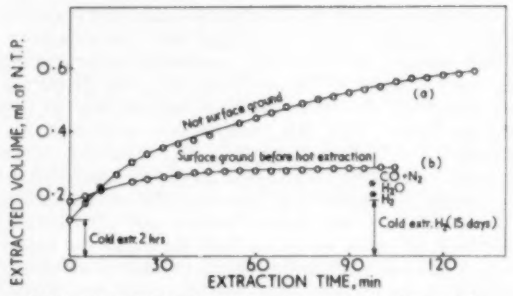
Whilst the information given in curves *a* and *b* explains the source of contamination qualitatively, comparison with Fig. 3 shows that the amount of carbon monoxide apparently varied from one test to another, possibly owing to variations in the extraction temperatures. A repeated extraction was therefore performed with the same sample at 740°C. Curve *c* confirms the expected effect of temperature. Again, it is parabolic with carbon monoxide as the main constituent.

Finally, the run-on and run-off pieces used in the preparation of the sample described were extracted at 740°C without any surface treatment. The parent plate was not the same as that used in the former tests; in particular, its nitrogen content was lower. In this



(*a*) Arrangement for assessment of longitudinal distribution
(*b*) Simplified arrangement for routine determinations
(*c*) Arrangement for assessment of transverse distribution

Fig. 1—Test-piece assemblies for hydrogen determinations

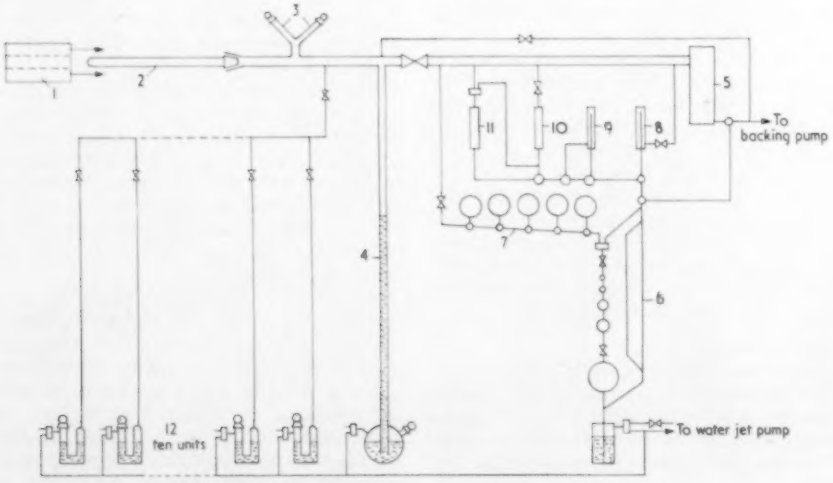


(*a*) Directly loaded sample (*b*) Sample ground before hot extraction

Fig. 3—Examples of gas evolution at 650–700 C

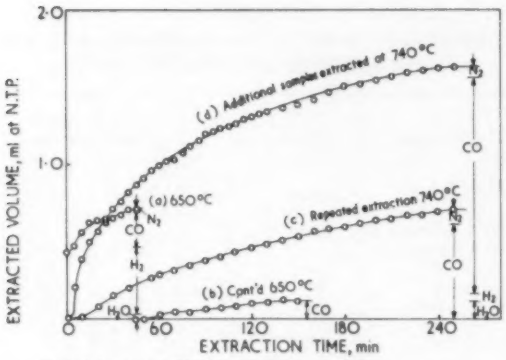
instance a substantial part of hydrogen was lost during storage at atmospheric pressure. For this reason, and because of differences in the size of sample, curve *d* should not be compared directly with the other curves in Fig. 4. It is included mainly as an illustration of the general course of gas evolution with tarnished samples directly exposed to high-temperature extraction. The steep slope of this curve over the first ten minutes cannot be explained by the evolution of H_2 and H_2O only, since these gases amounted to less than half the total volume obtained at that time. Thus the evolution of carbon monoxide is seen to proceed at a fairly high rate during the early stages of decarburization.

When hot extraction was performed at a temperature between 600° and 650°C there was hardly any effect of surface oxides. This is illustrated in Table II for two series of measurements with the arrangement shown in Fig. 1*a*. Although significant amounts of



(1) Furnace
(2) Quartz glass tube
(3) Compartments for used samples
(4) Mercury lift
(5) Three-stage diffusion pump
(6) Toepler McLeod gauge
(7) Storage volumes for measurement of instantaneous gas composition
(8) Palladium osmosis tube
(9) Copper-oxide furnace
(10) P_2O_5 drying tower
(11) CO_2 absorption tower
(12) Cold extraction chambers

Fig. 2—Apparatus for combined cold and hot extraction of hydrogen



(*a*) 21 h in vacuo at 20°C; hot extraction at 650°C
(*b*) Continued extraction of samples (*a*) at 650°C
(*c*) Repeated extraction of samples (*a*) at 740°C
(*d*) Quenched samples stored in air; extraction at 740°C

Fig. 4—Effect of time and temperature on gas evolution from tarnished samples

foreign gases were also present, a visible oxide layer did not cause any additional evolution of carbon monoxide.

It may be concluded from these tests that a chemical analysis is absolutely imperative in the extraction of directly-loaded samples at 740°C or higher, and is desirable also at lower extraction temperatures. Adsorbed moisture may cause some uncertainty, but absurdly high gas contents are not likely to arise on this account.

Calculated Rates of Gas Contamination

These results have been compared generally with the rate of gas evolution to be expected on theoretical grounds.* For carbon monoxide the measured values

* Details of the calculations and specific examples were given in an appendix to the original report, but are not included in this paper.

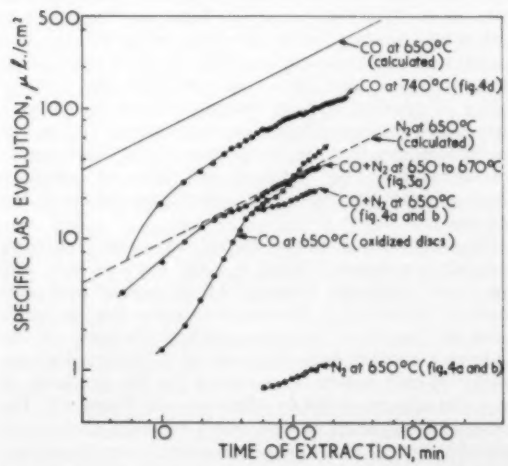

SPECIFIC GAS EVOLUTION, μL/cm²
TIME OF EXTRACTION, min
CO at 650°C (calculated)
CO at 740°C (fig. 4d)
N2 at 650°C (calculated)
CO+N2 at 650 to 670°C (fig. 3a)
CO+N2 at 650°C (fig. 4a and b)
CO at 650°C (oxidized discs)
N2 at 650°C (fig. 4a and b)

Fig. 5—Calculated and measured rates of gas evolution from oxidized samples of a steel containing 0·045% C and 0·01% N

Table I

CHEMICAL COMPOSITION OF GASES INDICATED IN FIG. 4

Curve	H_2O		H_2		CO		N_2		Total
	μl.	%	μl.	%	μl.	%	μl.	%	μl.
(a)	24	3·4	448	62·7	226	31·7	16	2·2	714
(b)	0	0	4	3·3	109	90·9	7	5·8	120
(c)	0	0	4	0·6	619	86·7	91	12·7	714
(d)	123	7·5	41	2·5	1415	86·9	50	3·1	1629

are about $\frac{1}{10}$ of the upper limit of the calculated rates; for nitrogen the factor is even less (Fig. 5). No adequate explanation for this can yet be given, and it will be interesting to discover if other types of steel show similar features.

Similar comparisons in relation to the effect of temperature on gas contamination show that the overall rate of decarburization at 650°C may be controlled by factors other than diffusion only.

Review of Earlier Determinations

Having established the main source of error in hot-extraction analysis, an attempt has been made to correct the values presented in the previous report. Samples of the electrodes used in 1955 were still in stock, although possibly with a different content of potential hydrogen in their coatings. Information on this point was available for one rutile electrode only. The coating of this electrode (R1X) had a lower content of moisture and total potential hydrogen after two years' storage. An additional sample of the same brand (R1) supplied in 1956 in a sealed tin had an intermediate moisture content. This batch had been stored in a closed container after breaking the seal, as were also the basic electrodes B 5 and B 6. Relevant data for all electrodes tested hitherto are given in Table III, together with the conditions of welding.

In Table IV a comparison has been made between the earlier results and the values obtained with the present improved technique. The correction to be applied for the basic deposits varies between 13 and 18 ml/100 g. Larger absolute corrections are required with the high-hydrogen deposits, in spite of the sharper end-point in extraction, and in spite of the

Table II

EFFECT OF SURFACE GRINDING ON TOTAL VOLUME EXTRACTED AT 600–650°C

Specimen and Position, X	Procedure	Gas Content, ml/100 g Deposit	
		Total Gas	Total Hydrogen
B 2 (X = 15 mm)	Cold extr.; ground	22·4	16·9
B 2 (X = 30 mm)	As welded	21·6	15·9
B 2 (X = 37·5 mm)	Cold extr.; ground	20·5	15·5
B 5 (X = 30 mm)	Cold extr.; ground	9·3	5·7
B 5 (X = 45 mm)	As welded	7·9	4·3
B 5 (X = 52·5 mm)	Cold extr.; ground	8·5	4·9

partial reduction of the oxide film caused by hydrogen in this instance. This rather unexpected behaviour is unexplained.

Electrodes R 1 and B 5 were tested simultaneously in the authors' laboratory and in Great Britain. The results and conclusions of these comparative tests are to be reported by the I.I.W.[2] Good agreement was obtained with beads from the basic electrode B 5 (±6% in terms of deposit and ±3% in terms of fused metal). A fair agreement was achieved with beads from the rutile electrode (±9 and ±20% in terms of deposit and fused metal, respectively). The greater dispersion of results with the high-hydrogen deposit is ascribed to a greater risk of hydrogen losses in this instance.

SPATIAL DISTRIBUTION OF HYDROGEN IN SINGLE-BEAD WELDMENTS

To the authors' knowledge no information has been published on macroscopic segregation of hydrogen in weld beads. The decreasing contents of hydrogen towards the end crater described by Meunier and Demarez[3] are related to variations in the potential hydrogen content of the electrode brought about during welding. Therefore this effect is not a result of a true segregation process.

The non-uniform distribution of hydrogen to be discussed probably developed during the cooling of a weld metal of initially uniform composition. It is believed that this particular kind of segregation is a unique property of dissolved hydrogen. Such concentration gradients are therefore not inconsistent with the commonly accepted view of a uniform concentration of other alloying elements in the axial direction of the bead.

Longitudinal Distribution of Hydrogen

The test-pieces and experimental details have already been described. Separate measurements of diffusible and residual hydrogen were performed on nearly all samples (*see* Table V). A few exceptions have been dealt with in Table II. The fused-metal weights have been reported to the second decimal place only, for they were computed from planimeter measurements of bead cross-sections, and not by precision weighing.

Table III

ELECTRODES AND WELDING CONDITIONS

Electrode	*I.I.W. Classification**	*Electrode Dia., mm*	*Welding Current, A*	*Welding Speed,† cm/min*
B1	E 355 B 10	4	175 (d.c.+)	18
B1X	E 355 B 10	5	230 (d.c.+)	13·5
B2	E 355 B 16	4	175 (d.c.+)	17
B3	E 254 B 26	4	175 (d.c.+)	16
B4	E 254 B 16	3·25	125 (d.c.+)	12
B5	E 455 B 11	4	175 (a.c.)	16
B6	E 355 B 10	4	175 (d.c.+)	13·5
R1	E 322 R 12	4	175 (a.c.)	11
R1X	E 322 R 12	4	175 (a.c.)	13·5
R2	E 342 R 22	4	170 (a.c.)	12
R3	E 333 R 11	4	170 (a.c.)	12
R4	E 334 R 12	4	170 (a.c.)	12
R5	E 132 R 12	4	170 (a.c.)	12
A1	E 254 A 12	4	175 (a.c.)	14·5

* Coding has been based on information given by the manufacturers.
† Approximately 1·5 in. of electrode is consumed per in. of bead.

Table IV

COMPARISON OF TOTAL GAS CONTENTS (1955) AND TOTAL HYDROGEN CONTENTS (1957)

Electrode	*Total Gas, ml/100 g (ref. 1)*		*Total Hydrogen ml/100 g (new det.)*	
	Deposit	*Fused Metal*	*Deposit*	*Fused Metal*
R1X (high-pot.H)	104	47	—	—
R1X (low-pot.H)	—	—	46	39
R1	—	—	53	33
R2	85	43	49	26
R3	76	40	52	20
R4	69	41	35	27
R5	55	35	37	25
B1	29·5	14·5	11·7	7·0
B2	29·0	10·0	16·0	8·1
B3	30·0	12·0	16·3	8·2
B4	32·0	18·5	13·5	9·5
B6*	—	—	9·5	6·5

* B6 is the same brand as B1.

The average holding times before quenching (col. 3 of Table V) are made up of two parts: the time required to finish the remaining part of the bead, and the delay from the completion of welding to the quenching of the weldment. The latter interval, of course, is constant within each series; the former is obtained from the position and the speed of welding. It is a matter of definition which reference point should be chosen for this calculation: the rear boundary of the end crater, the centre of the crater, or the end of the bead. It would seem reasonable to assign the same retention times to points that have solidified simultaneously. If the ripples on the bead may be taken as isothermal contour lines, the loop enclosing the end crater will delineate such points in the crater. This contour line is also better defined than the thermal centre of gravity, and the retention times in Table V have therefore been based on the distance from this rear boundary line to the centre of the slice under consideration. This method will lead to identical holding times for all slices containing parts of the end crater.

Diagrammatical presentations of total hydrogen contents are shown in Figs. 6 and 7 for one basic and one rutile electrode deposit. Much higher hydrogen contents prevailed in the crater samples than in other parts of the bead. A more gradual decrease of the hydrogen content was observed at a greater distance from the end crater. In contrast to the gradient of opposite sign reported by Meunier and Demarez[3], the present distribution was probably brought about by a loss of hydrogen through diffusion to the surroundings. This divergence of results may perhaps be attributed to two circumstances. First, the rather abrupt rise of the curves in Figs. 6 and 7 will not be revealed unless

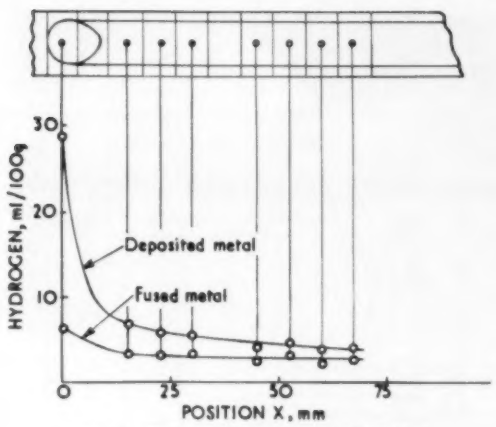


Fig. 6—Total hydrogen contents in B5 bead

sampling is continued into parts of the bead containing the end crater. Secondly, the very gentle slope of the curve observed at greater distance from the end crater might easily be overshadowed by a drying effect if the moisture of the electrode covering is readily expelled during welding. It is not known whether Meunier and Demarez obtained a series of moisture contents in their electrodes through a short exposure to humid atmospheres; in the present series, however, the electrodes were stored in fairly dry air for such a long period that parts of the moisture might have been tied up irreversibly in stable hydroxides.

The temperature range from freezing to about 1000°C is the one of importance for diffusion and, since this range will be traversed much faster in the end crater than in any other part of the bead, it is readily understood that dissolved hydrogen will be better preserved in, and close to, the end crater. Further indications of high-temperature diffusion as a controlling factor are shown in Table V. The concentration gradients, in terms of ml/100 g/mm distance, show little correlation with the delay between completion of welding and water quenching. Evidently the loss of hydrogen during a few seconds' delay after completion of welding is negligible in comparison with the losses occurring in the same time during welding. Finally, the very slight variations of hydrogen content observed at greater distances from the end crater are in accordance with the assumption that losses by diffusion occur mainly in the high-temperature range. Actually, there are in Table V a few instances of increasing hydrogen contents with increasing distance from the end crater. They are probably a result of scattering rather than of the drying effect mentioned.

The hydrogen contents reported for the deposit towards the end crater increase at a higher rate than the same data reported on the basis of fused metal. Again, this is mainly a crater effect. At a distance sufficiently remote from the crater the steady-state conditions would imply a constant ratio of deposited to fused metal. In the weld pool itself, however, liquid metal is being pushed by the arc in a direction opposite to that of welding. The ratio of deposit to fused metal will therefore usually be lower than the average when the pool suddenly solidifies as an end crater, although in manual welding this result probably depends largely upon the way in which the operator breaks the arc. In principle, however, it is possible to obtain end-crater samples that have been dug out by the arc so as to contain no deposit at all, or even a negative amount as determined by weighing. The fused-metal weight evidently is a better basis for a comparison between the average and the end-crater hydrogen contents.

Lateral Distribution of Hydrogen

The data of Table V indicate that high-temperature diffusion may be one of the factors controlling the longitudinal distribution of hydrogen in single beads. Since hydrogen can be transferred from a cooling bead to the surroundings only via paths that are normal to the axis of the bead, corresponding concentration gradients would be expected in the transverse direction. Such gradients would be a matter of considerable importance for a correct sampling and determination of the average hydrogen content.

An experimental verification of lateral gradients is far more difficult than the establishment of segregation in the axial direction. At present their existence has been demonstrated only in a few typical beads. For further confirmation, mid-plane and transverse sections of single beads were prepared for a metallographic examination in the following way.

Samples were obtained at liquid nitrogen temperatures by breaking apart the composite specimen shown in Fig. 1*c*. With the fracture facing downwards the specimen was clamped in a fixture that had also been pre-cooled in liquid nitrogen, and a plane section through the bead was obtained by intermittent grinding on metallographic emery paper immersed in a cooling bath at —80°C. Only a few strokes were taken at a time, between periods of cooling in liquid nitrogen. The preparation was finished with a light polish on a pre-cooled cloth impregnated with diamond dust. The evolution of hydrogen from surfaces prepared in this way was studied by means of the formation of gas bubbles in a film of oil applied to the surface at low

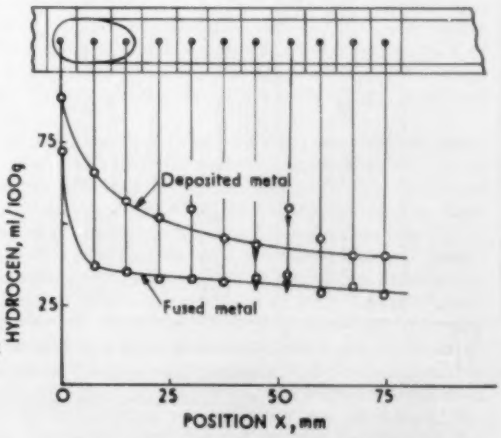


Fig. 7—Total hydrogen contents in R1 bead

Table V

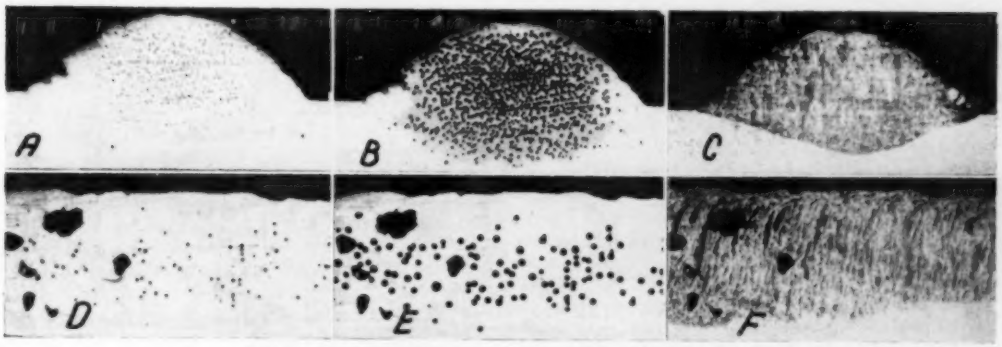
LONGITUDINAL DISTRIBUTION OF HYDROGEN

Electrode	*Sample Position X, mm*	*Time to Quenching, s*	*Deposit, g*	*Fused Metal, g*	*Hydrogen Contents, ml/100 g*			
					Diffusible		*Residual*	
					Deposit	*Fused Metal*	*Deposit*	*Fused Metal*
	0	3·4	0·645	1·13	12·8	7·3	2·9	1·7
	7·5	3·4	1·114	2·34	12·5	5·9	1·6	0·8
B1	15·0	3·7	1·245	2·15	10·4	6·0	2·4	1·4
	30·0	9·0	1·682	2·60	10·9	7·0	≥0·3*	≥0·2*
	37·5	11·6	1·541	2·46	11·5	7·2	2·7	1·7
	0	5·8	0·348	0·63	22·7	12·6	4·0	2·2
	7·5	5·8	0·817	1·82	17·4	7·8	2·2	1·0
B2	15·0	6·5	0·999	1·88	14·5	7·7	2·4	1·3
	30·0	11·9	0·966	1·93	13·4	6·7	2·5*	1·2*
	37·5	14·6	0·848	1·66	11·9	6·1	3·6	1·8
	0	4·3	0·185	0·68	40·0	10·9	6·0	1·6
	7·5	4·3	0·549	1·79	25·3	7·8	4·4	1·3
B3	15·0	4·9	1·048	2·13	14·8	7·3	2·5	1·2
	30·0	10·7	0·982	2·01	15·1	7·3	1·7	0·9
	37·5	13·6	1·087	1·93	13·4	7·6	1·7	0·9
	30·0	13·5	1·487	nd	11·6	nd	1·2	nd
B4	37·5	17·3	1·309	nd	13·1	nd	1·1	nd
	0	5·4	0·083	0·38	15·7	3·4	13·3	2·9
	15	7·3	0·756	1·49	5·2	2·6	1·6	0·8
	22·5	10·2	1·086	2·01	4·6	2·5	1·5	0·8
B5	30·0	13·1	1·181	1·91	4·9	3·0	0·8	0·5
	45·0	18·9	1·480	2·38	directly loaded		4·3†	2·7†
	52·5	21·7	1·253	1·91	4·1	2·7	0·8	0·4
	60·0	24·6	1·293	2·18	3·6	2·1	0·7	0·4
	67·5	27·5	1·355	2·12	2·9	1·8	1·5‡	1·0‡
	0	4·3	0·645	1·13	15·0	8·6	2·3‡	1·4‡
	7·5	4·3	1·147	1·98	9·8	5·7	1·2‡	0·7
B6	15·0	6·4	1·330	2·23	9·0	5·3	1·4	0·8
	30·0	13·2	1·712	2·27	7·7	5·8	0·9	0·7
	37·5	16·5	1·339	2·14	9·4	5·8	0.8‡	0·5‡
	0	3·0	0·186	0·23	81·1	65·5	8·6	7·0
	7·5	3·0	0·930	1·67	55·2	30·7	10·8‡	6·0‡
	15·0	3·0	1·206	1·94	49·3	30·6	8·0	5·0
	22·5	5·7	1·251	2·00	47·3	29·6	4·6	2·9
	30·0	9·7	1·345	2·21	50·2	30·5	4·8	2·9
R1	37·5	13·7	1·734	2·45	39·0	27·6	6·5	4·6
	45·0	17·8	1·861	2·53	≦39	≦29	5·3	3·9
	52·5	21·8	1·513	2·43	≦48	≦30	6·4	4·0
	60·0	25·8	1·574	2·49	38·3	24·2	7·9	5·0
	67·5	29·8	1·858	2·46	32·4	24·4	8·5	6·4
	75·0	33·9	1·586	2·26	30·8	21·6	9·7	6·8
	0	4	0·420	0·49	76·6	65·8	3·1	2·8
	7·5	4	0·946	1·59	48·6	28·9	6·2	3·7
R1X	15·0	5·6	1·218	1·85	43·7	28·7	4·9	3·2
	30·0	12·4	1·259	2·05	43·0	26·4	3·5	2·2
	37·5	15·8	1·254	1·96	40·0	25·6	3·5	2·2
	0	4·0	0·743	0·78	49·7	47·3	4·8	4·6
	7·5	4·0	1·048	1·88	47·9	26·7	7·8	4·3
A1	15·0	4·5	1·126	1·89	47·1	28·0	5·7	3·4
	30·0	10·7	1·317	2·15	40·2	24·6	8·4	5·1
	37·5	13·8	1·515	2·37	34·8	22·3	5·3	3·4

* Specimen not surface ground between cold and hot extraction.

† Including water vapour: 0·8 and 0·5 ml per 100 g deposit and fused metal respectively.

‡ Water vapour determined to less than 0·03 ml per 100 g deposit.



(*A*) After 70s in air at 20°C
(*B*) After 16 min in air at 20°C
(*C*) Fusion contour
(*D*) After 90s in air at 20°C
(*E*) After 375 s in air at 20°C
(*F*) Fusion contour

Note: Irregular black spots (*D*, *E* and *F*) are crevices not removed in grinding

Fig. 8—Evolution of gas from beads deposited with basic electrodes. Sections prepared at —80°C ×4

temperatures. At first the bubble formation was allowed to occur at controlled temperatures in a Dewar flask. Bubbles were observed above —5°C in basic electrode beads at about ×5 magnification. It soon became apparent that the same pattern of bubbles could be obtained with the specimen exposed to air at room temperature, and this method was used in subsequent work.

Typical bubble patterns are shown in Fig. 8 for sections through beads made with electrode B 5. No bubbles formed in a rim close to the surface, even after several minutes' exposure to room temperature. The width of this zone is about 1 mm in the mid-plane section, and about $\frac{1}{4}$ mm in the transverse section. This region was of course heated by the light required for photography, in the same way as other parts of the ground surface. In addition, the edges of the specimen were heated by convection and by conduction from the air, and hence these regions heated up faster than the central parts of the bead. If the presence of gas bubbles, and the time required for their formation may be taken as a measure of local hydrogen concentration, the lack of bubbles close to the surface of the bead may thus be regarded as a safe indication of a lower hydrogen content in this region.

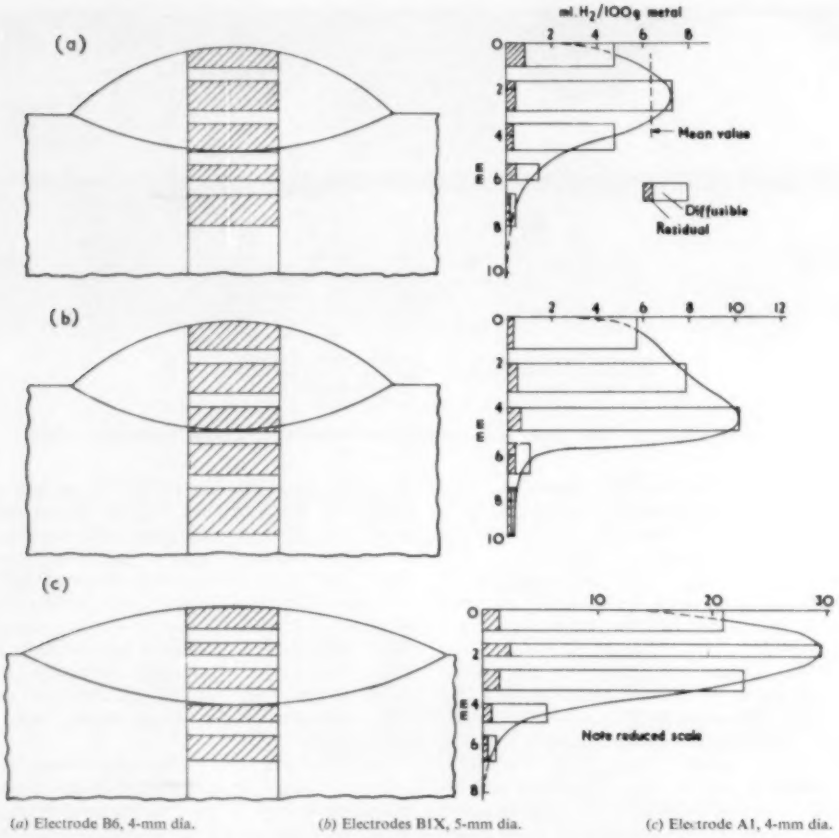
By comparison with the etched sections shown in the lower parts of Fig. 8 a less than average concentration of hydrogen is also indicated near the fusion line. This of course would be predicted from a mechanism of diffusion, although a loss of hydrogen from the bead to the parent-plate material would not be reflected in the analytical data of Table V.

It is not clear why the zone containing no bubbles is usually broader on mid-plane than on transverse sections. It is possible that the pattern of bubbles shown in Fig. 8 may also bear some relation to microporosity in the fused metal; however, weld-metal porosity cannot account for the evolution of gas from adjacent parts of the parent plate.

It is not possible at present to appraise the relative importance of local concentration and of porosity on the observed bubble pattern. An attempt has therefore been made to measure the hydrogen concentration at various levels by analytical means. The test-piece assembly shown in Fig. 1*c* was quenched in water, in CO_2–alcohol and in liquid nitrogen, immediately after the completion of welding. The weldment was broken apart at liquid nitrogen temperatures, and the specimen thus obtained was ground on its vertical fracture faces so as to permit a rigid clamping in a pre-cooled fixture. By cutting along the edge of this fixture with a jeweller's saw under a jet of liquid CO_2, slices about 1 mm thick were obtained at various levels in the bead and in the adjacent parent plate. These samples were analysed in two steps, as has been described earlier.

Results from tests performed on single beads from two basic and one acid electrode are shown diagrammatically in Fig. 9, the gas contents being reported in terms of fused metal or of parent-plate metal. These results confirm the trend predicted from the metallographic evidence of Fig. 8. (Note the different scale used for the acid electrode deposit.) This trend, and in particular the steep slope observed close to the surface of the bead, would be expected as a result of high-temperature diffusion. It is known that such concentration gradients develop during the early stages of isothermal diffusion.[4] In the present example this tendency would be enhanced because the outer layers of the bead were the last to cool. At the very surface of the bead, hydrogen would be able to escape at a stage when further transfer from the bulk of the bead had ended.

Significant amounts of hydrogen may diffuse from the weld pool into the parent plate; however, these amounts are completely recovered in the analytical determination. The effect of diffusion into the base material is illustrated in Figs. 8 *B* and *C*, and in Fig. 9, and is also well known from published data on underbead cracking. The presence of hydrogen outside the bead again raises the question of which basis should properly be used for reporting specific hydrogen contents. It has been shown that computation in terms of deposited metal may at times lead to unreasonable results for end-crater samples. For samples machined from all-weld metal, uncertainties are introduced because the result may depend on the location of the sample. In measurements performed on specimens



(*a*) Electrode B6, 4-mm dia. (*b*) Electrodes B1X, 5-mm dia. (*c*) Electrode A1, 4-mm dia.

Fig. 9—Lateral distribution of hydrogen in single beads

including parent-plate metal an unknown fraction of hydrogen does not belong to the weld metal, although of course it was originally present in the weld pool.

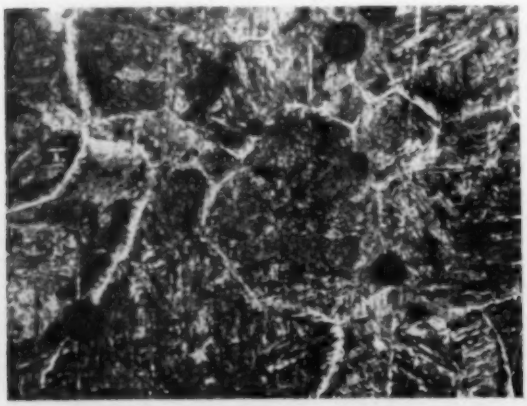


Fig. 10—Formation of bubbles from heat-affected zone of a steel containing 0·019% C and 0·44% Mn. Specimen prepared and etched at low temperature ×450

Apparently there is no single solution to this problem. For a metallurgical appraisal of the weld itself a fused-metal basis seems appropriate; for an appraisal of the electrode quality the deposited metal seems more satisfactory; and for questions related to the parent material the hydrogen should probably be reported in terms of unit length of bead.

Segregation of Hydrogen on a Microscopic scale

In contrast to the long-range concentration gradients resulting from diffusion at high temperatures, a segregation on a microscopic scale would be expected at considerably lower temperature in the $\gamma \rightarrow \alpha$ transformation. The abrupt drop of solubility connected with this transformation would tend to set up a high concentration in the austenitic grains close to the boundary of the growing grains of ferrite and, as has been discussed by Flanigan and Saperstein,[5] peak concentrations are likely to occur in grains of retained austenite. Such regions would be highly susceptible to microfissuring.

Whilst retained austenite may be an important factor in the internal cracking of alloyed steels, it is not likely to be present in sufficient amounts in a rutile-type weld metal containing about 0·06% C and a total

of about 0·40% Mn. The most probable location of microfissures in this instance is therefore the region of maximum hydrogen content, parallel and close to the columnar ferritic grains. Cracks in such positions are actually often observed, as has been illustrated in a previous report.[1]

In principle, the metallographic method described in the preceding section should lend itself to a study of microscopic segregation if the observations are interpreted with due caution. Thus, the vigorous evolution of gas observed from the cracks in a quenched bead should be taken as an indication of a greatly increased surface in the crack itself, rather than of a high local concentration only. Similar errors may arise in the study of essentially crack-free basic electrode deposits through the presence of microporosity. It appears therefore that the method could be tested more safely in the parent plate close to the fusion line, where such sources of error are excluded.

A fairly typical pattern of gas bubbles is shown in Fig. 10. With few exceptions, the bubbles are located at or close to the seams of ferrite delineating the boundaries of formerly austenitic grains.

The usual precautions were taken in the preparation of this specimen to reduce the risk of levelling out concentration gradients before the test. Some redistribution of hydrogen may have taken place during the time required for application of oil to the specimen and for adjustment of the microscope. It has been noticed repeatedly, however, that a preferential evolution of gas from the grain-boundary ferrite will persist for a long time. When the bubbles already formed are wiped away, it is usually found that new ones are formed at a higher rate in, and close to, the ferrite seams than in other positions. This may perhaps be interpreted as an indication of factors other than local concentration contributing to the observed pattern of bubbles. Because of a greater mobility of hydrogen in the ferritic envelopes than in the regions containing precipitated carbides, the former may perhaps serve as drainage in the same way as a crack. More experimental information is required for an appraisal of the relative importance of high local concentrations and of the draining effect in the formation of the bubble pattern shown in Fig. 10.

CONCLUDING REMARKS

Various aspects of the analytical procedures and equipment have been discussed in the first part of this report, and it may be sufficient to state that separate measurements of 'diffusible' and 'residual' hydrogen seem preferable to a determination of both in a single step. It should be emphasized that these rather loosely defined terms have been used merely for the sake of convenience. They are both functions of the experimental conditions; in particular, the contents of 'diffusible' hydrogen determined with the present arrangement should not be compared directly to, e.g. the hydrogen content defined by the A.S.T.M. method.

A further restriction in the use of the data reported in this investigation has been demonstrated in the studies of the longitudinal distribution of hydrogen. It is evident that a sample located midway between the start and end points of a bead, however drastically quenched, can at the best give a conventional figure. The hydrogen contents determined in such samples can, of course, be used for a comparison of electrodes, and of drying procedures and other factors of practical importance. The hydrogen content of the weld pool on the other hand will be more correctly represented by samples taken from the end crater. Thus the apparent equilibrium temperatures computed by Mallett and Rieppel[7] from solubility data and from their measurements of hydrogen contents would probably be displaced towards higher values if such tests were repeated with samples taken from the end craters.

The consequences of the transversel segregation of hydrogen in regard to sampling techniques have been shown to depend on the purpose of the measurement. The authors consider that the usual practice of reporting the hydrogen content in terms of deposited metal is more convenient in most applications; however, useful information may be obtained by an additional determination of the deposit/fused-metal ratio.

ACKNOWLEDGMENTS

The authors are indebted to members of SINTEF's metallurgical staff for their assistance in the experimental part of this investigation, in particular to Miss V. Krogstad and to Mr. R. Gravem. They also desire to thank members of the Hydrogen Sub-Committee of the International Institute of Welding for counsel and advice.

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Gas Porosity and Sources of Hydrogen in the Metal-arc Welding of Light Alloys

By W. G. Hull, F.I.M., and D. F. Adams, A.I.M.

SYNOPSIS

THE experimental work described, has shown that hydrogen may be derived from a number of sources within the electrode including free moisture, water of crystallization, and corrosion product at the core-wire surface, and that each individually can contribute towards weld metal porosity. Although moisture can be removed by baking, it would appear that any significant improvement in soundness hinges on the development of fluxes which are free from water of crystallization and also on the prevention of core-wire corrosion by surface protection or control of storage conditions.

Introduction

THE work reported here on weld porosity forms part of a programme being carried out to investigate various problems encountered in the metal-arc welding of H.10 alloy. Work on other aspects namely dilution[1] and weld cracking[2] has been reported earlier.

In recent years, there has been a revival of interest in the process in certain fields of light alloy fabrication and although there have been noticeable improvements in the performance of electrodes currently available, the general level of weld quality obtained is still considerably below that for joints made with argon-shielded processes. Published work dealing directly with the causes of porosity in metal-arc welds is scanty, the general view being that the hydrogen responsible for the unsoundness is derived from dissociation of free moisture present in the flux covering. Tests carried out earlier by the authors had suggested, however, that other sources of hydrogen might also be available in the electrode and of possible importance in relation to porosity.

The object of the work described has been, therefore, to examine the effect of free moisture in more detail and also to establish the identity and relative importance of other factors likely to lead to the development of porosity in metal-arc welds.

PRELIMINARY WORK

Initially, it was desirable to have some indication of the extent of porosity in metal-arc welds made under industrial workshop conditions. Quantitative estimates of the porosity were made in a number of commercial welds employed in the fabrication of a small structure in the H.10 alloy. The estimates were done by the density method described later. The various welds examined in this way gave porosity values within the range 3·0–4·7% voids. The size and distribution of the pores in one of the specimens estimated to contain 3·8% voids is shown in Fig. 1.

Direct evidence that the electrode itself was instrumental in introducing porosity into the joint was obtained by metallographic examination of the fused ends of electrodes used for various welds made in the laboratory. The microstructure at the melted end almost invariably exhibited marked pinhole porosity, suggesting that the metal being transferred had a high gas content before traversing the arc gap. A photomicrograph illustrating stub-end porosity is shown in Fig. 2, which also includes a microsection prepared from the end of the corresponding weld. Another photograph, Fig. 3, which is a longitudinal section through a weld crater, illustrates the transfer of unsound material from the core wire (shown at the top) to the weld pool during short circuiting. This specimen was obtained by interrupting the current suddenly while welding was in progress.

SCOPE OF INVESTIGATION

It was assumed beforehand that the hydrogen derived from the electrode originated mainly from free moisture present, but the possibility that other sources of gas were also available was considered. These it was felt might include the following:

(*a*) Water of crystallization in the flux
(*b*) Hydrated oxide or corrosion product at the core wire surface
(*c*) Internal hydrogen of core wire material.

It was decided therefore to obtain information on the following aspects to assess their importance in relation to weld unsoundness.

Free Moisture

(i) Levels of moisture in 'as-received' electrodes
(ii) Suitable drying treatments
(iii) Moisture re-absorption on exposure after drying
(iv) Moisture loss while welding is in progress.

Water of Crystallization

(i) Amounts present in typical electrode coatings
(ii) Removal during the above drying treatments.

Core-wire Hydrogen

(i) Amounts available from surface
(ii) Amounts held internally.

MATERIALS

Electrodes

Electrodes having an aluminium–10% silicon core wire were used for this part of the investigation. These were obtained from two manufacturers *A* and *B* and

Report *LM* 1/57/58 of the Association issued to members in February, 1958.

Mr. Hull is Head of the Metallurgical Laboratories, Atomic Power Constructions Ltd., Heston, and Mr. Adams is a Research Officer with the *B.W.R.A.*

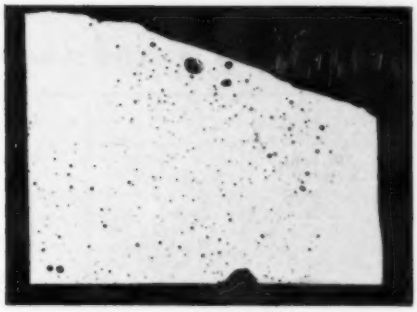

Fig. 1*a*—Commercial weld containing 3·8% voids ×3½

were assumed to be reasonably representative of the type of electrode normally supplied to industry. Analyses of random samples showed that the core-wire composition was to specification NW2, i.e. 10–13% silicon.

Plate

Welding tests were carried out on ¼-in. thick plate to specification H10 WP the actual composition being as follows:

Cu, %	Mg, %	Si, %	Mn, %	Fe, %
0·02	1·14	0·97	0·74	0·28

EXPERIMENTAL PROCEDURE

The method of sampling and testing each electrode condition investigated is shown diagrammatically in Fig. 4. For the 'as-received' condition, the electrodes were removed from a previously unopened packet and sampling and welding carried out with the minimum of delay. Where drying was involved, an electrode of each type was loaded in an electrically heated air circulation furnace and baked for one hour at temperature, after which the electrodes were removed and allowed to cool to approximately room temperature. The sample end was then sent for gas analysis and moisture determination, the latter being commenced immediately. Simultaneously with the sampling, the welding test which is described later was made with the remainder of the electrode. A sample of coating on the stub end was also sent for free-moisture analysis to obtain data on the loss which had occurred during welding.

The first batch of tests related to conditions where virtually no exposure to the atmosphere had occurred, the electrode conditions being 'as-received' and dried at 110°, 150°, 200°, and 300°C. A second series was also run in which electrodes with the above pre-treatments had been exposed to the atmosphere for a period of twenty-four hours. The relative humidity over the final six hours was recorded. The methods used for free moisture and hydrogen determination are given in Appendixes I and II.

The welding test consisted of a single-pass butt joint in 6 in. × 4½ in. × ¼ in. thick HP10WP plate having a square edge preparation. The welds were laid over a backing bar at an overall speed of 12–15 in./min, the electrode operating at nominally 210 or 240A, depending on the type being used. Before welding, the test plates were de-greased and carefully cleaned by scratch brushing.

An estimate of the percentage porosity in the weld metal was obtained from a density specimen taken from the centre 2 in. of each run. The determination was made by weighing the density specimen in air and in carbon tetrachloride using an automatic semi-micro balance accurate to 0·1 mg. Carbon tetrachloride was used in preference to water since its superior wetting properties tend to minimize errors due to cavities in the specimen surface.

The percentage voids were calculated as follows:

$$\% \text{ voids} = \frac{(d_1 - d_2)\,100}{d_1}$$

Where d_1 = density of sound material
d_2 = density of weld metal.

A value for the density of sound material of weld metal composition was obtained from specimens machined from a small porosity-free ingot. The composition of the ingot was based on the average composition of several weld metal deposits. The alloy was melted in a plumbago crucible by argon arc, which prevented oxidation and also helped to de-gas the melt by the stirring action provided.

Casting was carried out in a cylindrical steel mould also under argon. The ingot analysis obtained was reasonably representative of average weld metal composition as shown below:

	Mg, %	Si, %	Fe, %
Ingot	0·8	2·2	0·5
Weld metal (average)	0·6	2·5	0·3

Specimens were obtained from the centre of the casting which was shown to be sound by microscopical examination. By this method, a density value of 2·71

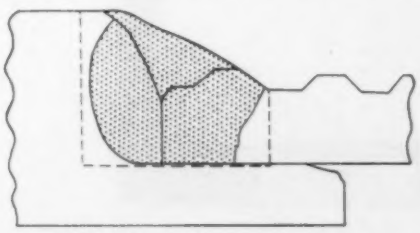

Fig. 1*b*—Details of joint shown in Fig. 1*a*

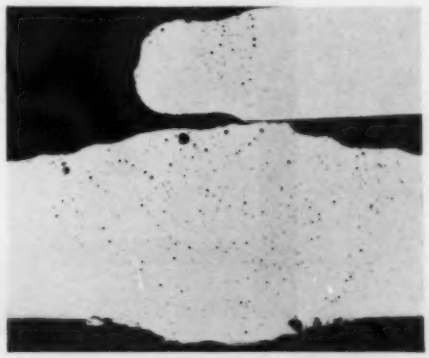

Fig. 2—Porosity in electrode-end and in corresponding weld ×3½

Table I

DATA RELATING TO FREE MOISTURE CONTENT OF ELECTRODE COATINGS AND WELD POROSITY

Electrode type and number	*Condition*	*Exposure, h*	*Humidity, %*	*Moisture, %*		*Weld, % voids*
				*Before welding**	*After welding†*	
A1	as received	nil	—	0·28	0·17	1·2
A2	,,	24	73	0·83	0·14	2·5
A3	dried 110°C	nil	—	0·21	0·09	2·6
A4	,,	24	74	0·91	0·27	2·3
A5	dried 150°C	nil	—	0·08	0·08	3·1
A6	,,	24	79	2·34	0·88	3·4
A7	dried 200°C	nil	—	0·01	nil	2·8
A8	,,	24	72	0·40	0·06	2·7
A9	dried 250°C	nil	—	0·11	0·08	3·1
A10	,,	24	75	0·53	0·08	1·2
A11	dried 300°C	nil	—	0·13	0·06	2·0
A12	,,	24	75	0·44	0·17	2·7
B1	as received	nil	—	0·40	0·21	1·0
B2	,,	24	73	0·55	0·18	1·2
B3	dried 110°C	nil	—	0·09	0·05	1·2
B4	,,	24	74	0·55	0·15	0·4
B5	dried 150°C	nil	—	0·03	0·03	1·2
B6	,,	24	79	1·23	0·21	2·1
B7	dried 200°C	nil	—	0·01	0·01	1·6
B8	,,	24	72	0·48	0·10	1·8
B9	dried 250°C	nil	—	0·05	0·02	1·3
B10	,,	24	75	0·48	0·16	0·4
B11	dried 300°C	nil	—	0·07	0·03	—
B12	,,	24	75	0·42	0·21	1·1

* In front-end sample † In stub end

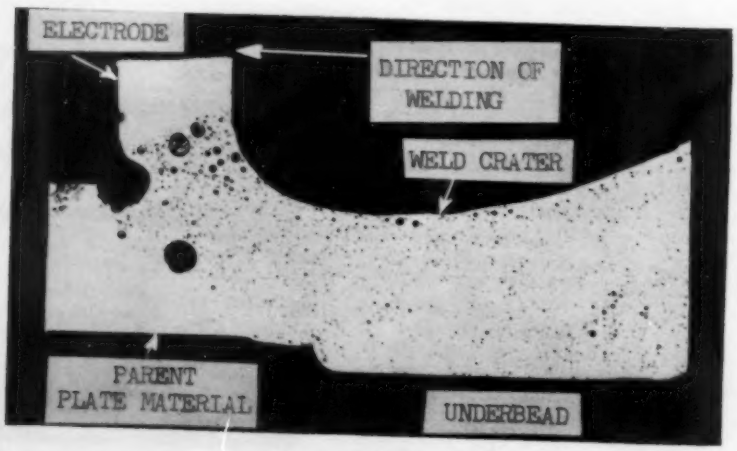

Fig. 3—Longitudinal section through weld crater showing introduction of porosity from the electrode during short circuiting ×3½

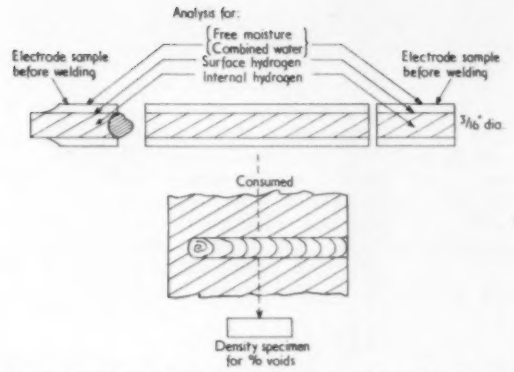

Fig. 4—Sampling and testing procedure to evaluate the effect of electrode condition on weld porosity

was obtained and the percentage voids quoted for weld metal is based on this figure.

RESULTS

Free Moisture in Coating

Effect of drying—The free moisture contents of electrode coatings in the various specified conditions are given in Table I. It will be seen that coatings of electrodes taken from a previously unopened packet had a certain amount of moisture present, i.e. 0·28% and 0·40% for electrodes *A* and *B* respectively even though drying had been presumably carried out by the manufacturers before packing.

Table II

WATER OF CRYSTALLIZATION IN ELECTRODE COATINGS

Electrode type and number	*Electrode condition*	*Water of crystallization*		*Weld, % voids*
		Before welding	*After welding*	
*A*1	as received, used immediately	0·58	0·61	1·2
*A*7	dried 200°C, used immediately	0·55	0·55	2·8
*A*11	dried 300°C, used immediately	0·45	0·48	1·0
*B*1	as received, used immediately	0·42	0·42	1·6
*B*11	dried 300°C, used immediately	0·40	0·44	N.D.

N.D. = No determination

Table III

INTERNAL AND SURFACE HYDROGEN OF CORE WIRE

Electrode type and number	*Condition*	*Sample location*	*Hydrogen, ml/100 g*			*Weld, % voids*
			Internal	*Surface*	*Total*	
*A*2	as received exposed 24 h	start*	—	—	6·2	2·5
		stub end†	—	—	7·3	
*A*3	dried 110°C used immediately	start	—	—	10·2	2·6
		stub end	—	—	3·8	
*A*4	dried 110°C, exposed 24 h	start	—	—	6·8	2·3
		stub end	—	—	7·7	
*A*7	dried 200°C, used immediately	start	0·7	4·3	5·0	2·8
		stub end	0·7	5·1	5·8	
*A*11	dried 300°C, used immediately	start	0·7	5·5	6·2	2·0
		stub end	0·5	3·8	4·3	
*B*1	as received, used immediately	start	—	—	—	1·0
		stub end	—	—	2·7	
*B*2	as received, exposed 24 h	start	—	—	6·6	1·2
		stub end	—	—	3·3	
*B*7	dried 200°C, used immediately	start	0·2	1·6	1·8	1·6
		stub end	0·4	3·6	4·0	

* Sample from front end of electrode before welding. † Sample from stub end of electrode after welding

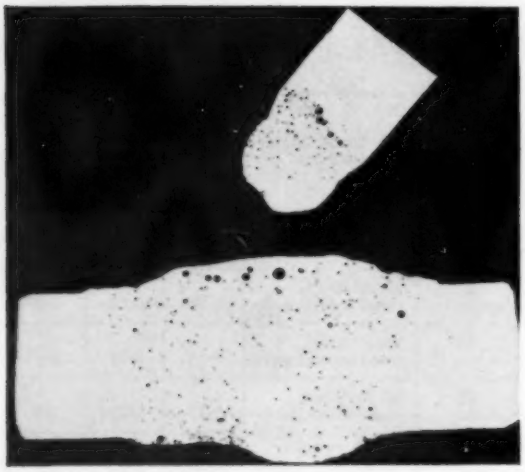

Fig. 5—Porosity in weld made with a hot type *B* electrode (300°C) ×3½

A reduction in moisture content occurred at each of the drying temperatures used, a minimum level being obtained by drying at 200°C. Treatments at temperatures exceeding 200°C were apparently less effective in removing free moisture but the results obtained were probably affected either by the slight time-lag in sampling or because a small moisture pick-up had occurred during cooling to room temperature. Since sampling was carried out in a matter of seconds after cooling, the results demonstrate the marked sensitivity of the coatings to moisture absorption immediately after drying.

Effect of electrode exposure after drying—The extent of moisture pick-up by the dried coating on subsequent exposure to the atmosphere is shown also in Table I. A time of twenty-four hours was chosen, since it was felt that such an extended period might conceivably occur in practice where the work was interrupted and the electrodes not returned for drying.

The results show that although a rise in moisture level invariably occurred, the amount, as expected, depended on the relative humidity prevailing at the time and was most marked where the humidity was 79.%

Moisture losses during welding—It would be expected that where free moisture was present in an electrode coating, there would be a progressive decrease as the electrode was consumed due to the temperature rise in the core wire. To confirm this and to assess the extent of the loss, moisture analyses were obtained from the stub ends immediately after welding. Table I gives the results obtained and shows that a reduction in moisture invariably occurred but that where an appreciable amount was originally present moisture removal was incomplete. In general the moisture level in the stub end is shown to fall to a value approximately 10–30% of that originally present.

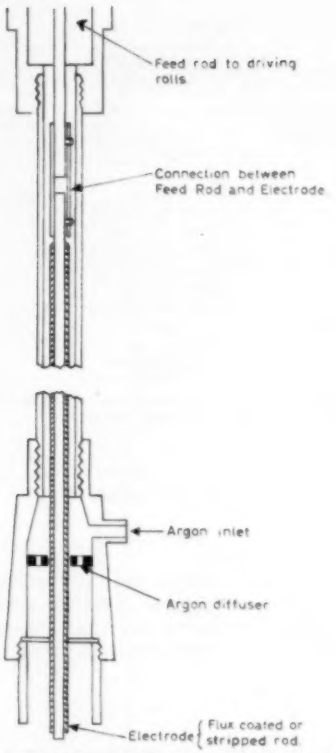

Fig. 7—Extension for experimental self-adjusting arc head

Water of Crystallization in the Coating

Both types of coating were found on analysis to contain water of crystallization amounting to about 0·5%. The results are included in Table II. It will be seen that no reduction in the amount of combined water in the coatings had occurred either by drying at temperatures up to 300°C or by the heat developed in the core wire during welding.

Hydrogen associated with the Core Wire

Internal hydrogen—Samples of core wire taken from several of the electrodes used in the investigation gave internal hydrogen figures mainly within the range 0·5–0·9 ml/100 g although on one or two samples results as low as 0·2–0·4 ml/100 g were obtained. The results are included in Table III. The variations in the figures obtained were not unexpected since there was no guarantee that the core wires in each packet originated from the same coil of wire.

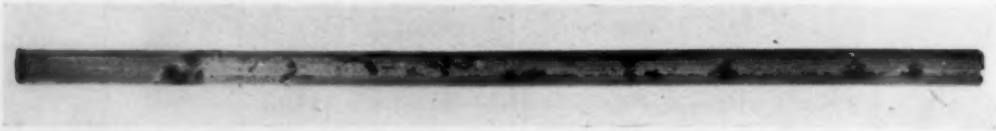

Fig. 6—Corroded surface of electrode core wire

Surface hydrogen—The analysis results are given in Table III and show that, compared with the internal hydrogen figures, relatively large amounts of gas could be derived from the surface of the wire. Examination of the surface condition of the core wires tested revealed substantial amounts of corrosion product and it is assumed that the gas was derived mainly from this source. The surface appearance of a core wire from one of the electrode packets used is shown in Fig. 6. The gas analysis results show that there was often considerable variation in the hydrogen evolved from the two samples selected from the electrode. This can be explained by the intermittent distribution of the corrosion sites.

Weld Porosity

The analyses results have shown that besides free moisture there were a number of other sources of gas available, namely water of crystallization in the flux and the hydrogen associated with core wire. Examination of the welds made with these various electrodes indicated that one or all of the additional sources were active in introducing unsoundness. Thus in Table I, electrodes used immediately after drying and shown by analysis to be virtually moisture-free all produced porous weld metal containing in certain cases as much as 3·0% voids.

The results were somewhat contrary to expectation at the time and the small moisture pick-up thought to have occurred prior to welding was considered as a possible explanation for this apparent anomaly. To investigate this, test welds were made with electrodes baked at 200° and 300°C and used straight from the furnace before any appreciable loss in temperature had occurred. It was assumed that under these conditions the free moisture content of the coating would be nil. Subsequent porosity measurements showed the welds to contain 2·5–3·5% voids; a micro-section of one of the welds together with the fused end of the corresponding electrode is shown in Fig. 5.

It was therefore concluded from these tests that the other known sources of hydrogen were important in relation to porosity but it was not possible at this stage to evaluate their effect individually. It was felt that the internal gas content of the core wire was unlikely to be a major factor, since only relatively small amounts were present, compared with the hydrogen available from the core wire surface and that which could be derived from the water of crystallization.

Further tests designed to assess the individual effects of the various sources of hydrogen existing in metal-arc electrodes were subsequently carried out and are dealt with in the following section of the report.

POROSITY IN WELDS MADE WITH SPECIALLY PREPARED ELECTRODES

Information was obtained from the work just described of the various sources of hydrogen present in commercial electrodes but no positive conclusions could be drawn as to their relative importance in introducing weld porosity. In an attempt to evaluate the individual effect of these various sources, further welding tests were carried out using specially prepared core wire and coated electrodes. The composition of the core wire and plate materials was the same as before i.e. Al–10% Si and HP10WP. The tests were made with the following types of electrodes:

(i) Carefully cleaned bare wire to assess the effect of internal hydrogen
(ii) Corroded electrode core wire with flux coating removed to indicate effect of surface hydrogen
(iii) Dry flux-coated electrode, freshly coated on clean bare wire to obtain data on effect of water of crystallization
(iv) As above, but with coating undried for effect of free moisture.

Experimental Procedure

Since it was essential to employ argon shielding for tests made with bare wire, careful consideration was given to the welding method to be used. It was desirable for the mode of metal transfer to be similar in character to that in metal-arc welding and for this reason the self-adjusting arc process was selected. The controlled arc method would probably have been more suitable but the equipment was not available.

Exploratory tests were carried out with the B.W.R.A. self-adjusting arc unit (described elsewhere[3]) to establish operating conditions to give the short-circuiting globular type of transfer typical of metal-arc welding. Suitable conditions were obtained by running the arc from 10 s.w.g. wire at well below the threshold current to give the globular transfer required and at a very short arc length to induce frequent short circuiting. The current used was 200–220 A giving a burn-off rate of 55–60 in./min.

To avoid damage to the fluxed or corroded surface of the electrode under test, a nozzle extension was fitted to the torch to enable the electrode to be fed without passing through the feed rolls. In this arrangement, which is shown in Fig. 7, the electrode was fed to the arc via a feed rod driven by the rolls in the head of the torch. Argon was supplied to the nozzle assembly at the end of the extension tube passing through a perforated disc to provide a reasonably uniform gas flow. A general view of the equipment used is shown in Fig. 8. Welding was carried out automatically with the plate traversing at 17–20 in./min.

For tests made with clean bare wire, extreme care was taken to ensure that the wire surface was as free from contamination as possible so that any unsoundness in the deposit would be associated with the internal hydrogen in the material. The wire was first de-greased, acid pickled, and then mechanically cleaned with steel wool and the procedure repeated immediately prior to the test being carried out.

In preparing wire for tests to indicate the effect of surface hydrogen, the flux was detached and any residual flux particles carefully removed. By this means the existing surface condition of the core wire was retained as far as possible although a certain proportion of the corrosion product was unavoidably removed with the flux.

The covered electrodes used were prepared and tested in collaboration with an electrode manufacturer. The operations involving core wire surface preparation, application of the coating, electrode drying, and welding the test plates were completed on the same day to minimize the possibility of the results being affected by core-wire corrosion. The core wire was cleaned as mentioned previously just prior to

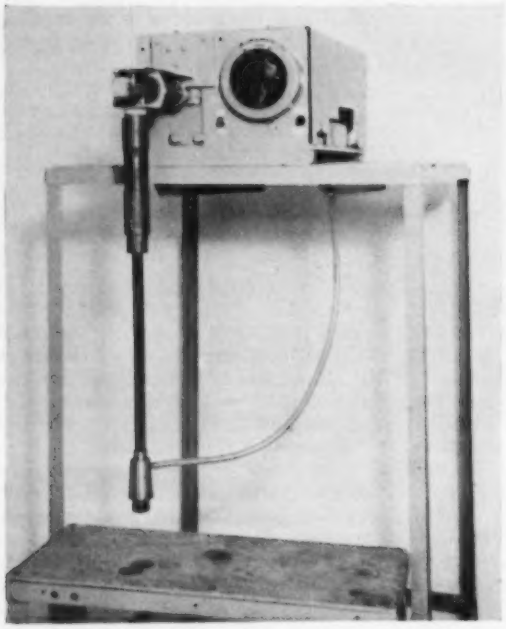

Fig. 8—B.W.R.A. self-adjusting arc unit with feed-tube extension

coating after which the electrodes were furnace heated at 250°C for ten minutes to drive off the majority of the moisture. For tests in the moisture-free condition, sufficient electrodes were further dried for one hour at 200°C and the deposits made with the electrodes still hot as described earlier in the report. The remaining electrodes which had been only partially dried were used for tests to indicate the effect of free moisture.

Results

Clean bare wire—The weld deposits made with this type of electrode showed very little unsoundness, the highest porosity figure obtained in the microsections examined being 0·3% voids. A section prepared from one of the welds is shown in Fig. 9 with the corresponding electrode end in which only a trace of porosity is visible.

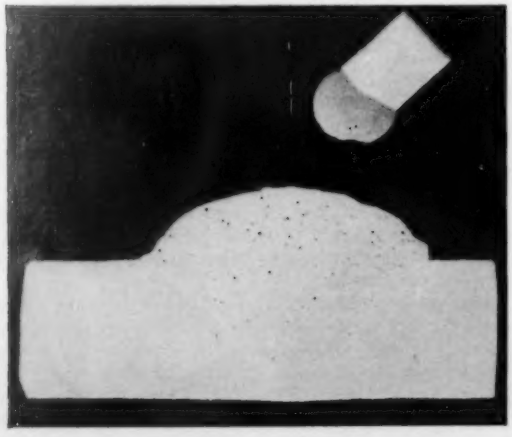

Fig. 9—Weld metal deposited from clean bare wire ×3½

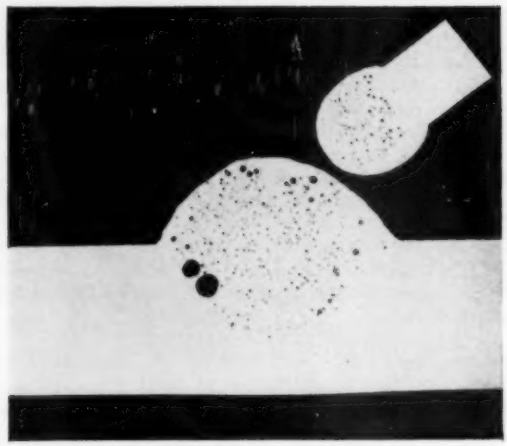

Fig. 10—Weld metal deposited from metal-arc electrode core wire with flux removed ×3½

Although the assumption that the porosity was derived from the gas present internally in the wire may be open to doubt, the results provide sufficient evidence to show that, under conditions of extreme surface cleanliness, virtually sound weld metal can be produced.

Corroded electrode core wire—Weld metal produced by bare core wire corroded by the coating during storage exhibited unsoundness approaching that often obtained in metal-arc welding, i.e. 2·0–4·0% voids. A photomicrograph showing an example of the type of unsoundness encountered is shown in Fig. 10, which also illustrates the porosity which had developed in the fused end of the wire used.

The results of these tests confirm the suggestion given by earlier work namely that appreciable

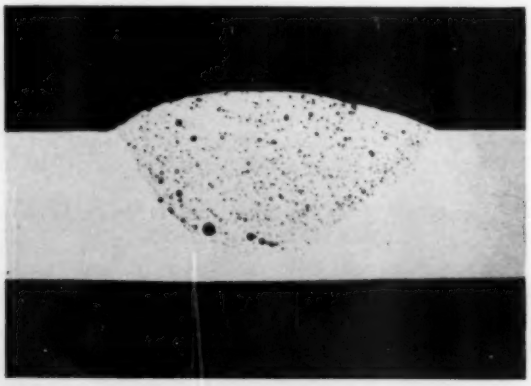

Fig. 11—Porosity in weld made with freshly coated and dried electrode using clean core wire ×3½

unsoundness can develop where the core-wire surface is contaminated with corrosion product.

Dried freshly covered electrodes—Examination of welds made with electrodes in this condition, namely with only internal hydrogen and water of crystallization present, exhibited porosity within the range 2·0–3·0% voids as illustrated in Fig. 11. Since it was shown that only slight porosity is derived from the internal hydrogen in the core wire, the results indicate that the combined water in the coating can be an active source of weld unsoundness.

Moist freshly covered electrodes—Welds made with new electrodes containing moisture in the coating in addition to water of crystallization exhibited rather more porosity than those produced from dry electrodes.

It so happened that there was considerably more free moisture in these coatings than is normally the case. Nevertheless the increase in porosity is assumed to be due to this additional source of hydrogen.

DISCUSSION

Several facts, having a bearing on the problem of unsoundness in metal-arc welds, have emerged from the results of the work carried out. Firstly, under conditions of reasonable plate cleanliness, the porosity normally encountered is derived almost entirely from hydrogen associated with the electrode. Although no attempt has been made to elucidate the mechanism governing the introduction of gas to the weld pool, there has been ample evidence from metallographic examination of the fused ends of core wire that the molten material has a substantial amount of gas in solution before entering the pool. Furthermore, visual observation of the arc suggests that quite a high proportion of the metal is delivered directly to the pool during the frequent short circuiting, which occurs with the arc lengths normally used. As with other consumable electrode processes such as self-adjusting arc, a reduction or elimination of short circuiting would be expected to minimize the weld metal porosity and a few tests recently carried out with metal-arc electrodes using longer arc lengths have tended to confirm this. It is unlikely, however, that this would be a practical method for improving weld metal soundness since the process being essentially for manual operation does not lend itself to any precise control of arc length.

In practice, better weld quality hinges on the development of suitable methods for eliminating the various sources of hydrogen found to be present in the electrodes, such as water of crystallization, free moisture, and surface contamination of the core wire. With the coatings of the electrodes investigated no loss in the amount of water of crystallization present was obtained by baking at temperatures up to 300°C, which suggests a modification of coating composition as the only means of removing this source.

Free moisture in the coating at the time of welding can also lead to porosity but, although it is effectively eliminated by baking, re-absorption quickly occurs on exposure to the atmosphere after cooling. A possible practical method for overcoming this, as adopted in the work here, would be to use the electrodes straight from the drying oven whilst still hot. Another, but less convenient procedure would involve the storage of the dried electrodes under completely moisture free conditions.

The surface condition of the core wire has also been shown to have an important bearing on the incidence of porosity. With the electrodes used, which were only a few weeks old, there was considerable contamination of the surface from corrosion by the flux during storage. Analyses have shown that a surface condition of this type is a potentially rich source of hydrogen which in welding can lead to gross unsoundness.

Although corrosion of core wire in contact with the type of fluxes used would be difficult to prevent, the presence of free moisture would be expected to accelerate the rate of attack during the storage period. In this connection both types of electrodes used were found to contain free moisture when taken from a previously unopened packet and it is presumed that this was due either to inadequate drying after manufacture or to absorption from the packaging materials employed.

Other factors likely to influence the course of corrosive attack would include the surface condition of the core wire prior to coating and obviously the storage period. Work on these aspects has recently been carried out and will be reported in the near future. Although it is considered here that corrosion product is probably the main source of hydrogen contributed by the core wire, other surface contaminants such as a heavy hydrated oxide layer, formed during working and annealing, may well be important in relation to porosity, as is the case with the self-adjusting arc process, but would only assume importance if corrosion effects could be successfully inhibited.

CONCLUSIONS

The following main conclusions may be drawn with reference to the particular commercial electrodes used:

(1) The porosity exhibited by the welds examined was introduced by hydrogen derived from the electrodes.

(2) Three main sources of hydrogen were identified, all of which could produce significant amounts of porosity in the weld metal. The sources referred to were:

(*a*) Water of crystallization in the coating
(*b*) Free moisture in the coating
(*c*) Contamination of the core wire surface by corrosion products due to attack by the coating.

(3) Any major improvement in the soundness of welds made by this process will depend largely on the extent to which all the above sources of hydrogen can be excluded.

(4) In the electrodes used, water of crystallization was not removed by baking at temperatures up to 300°C. The elimination of this source would therefore involve the development of coatings having no water of crystallization or alternatively having constituents which lose combined water at relatively low temperatures, viz.: 300°C.

(5) Free moisture could be removed by baking at temperatures up to 300°C but reabsorption by the dried coating readily occurred on exposure to the atmosphere.

(6) Various factors will influence the rate of corrosive attack of the core wire by the coating and the amount of hydrated corrosion product formed. Of these, free moisture in the coating during storage, the surface condition of the wire, and the type of coating composition used are likely to be important.

APPENDIX I

Methods used for Moisture and Water of Crystallization Determinations in Coatings

By J. Walker, L.I.M.

Free Moisture

To determine the free moisture content, the samples of coatings were removed from the welding rods, and without further treatment were dried to constant weight at 110°C in a laboratory air oven.

Water of Crystallization

The technique employed was based on a method by Gayley and Wooding[4]. The method consists of heating the sample to 1300°C in a stream of dry oxygen and collecting the water in a desiccating train and weighing. To obtain low blank values the combustion boats were pre-heated at 1000°C and then at 1300°C in the combustion furnace in a stream of dry oxygen overnight. The furnace was run continuously for 1300°C for the duration of these tests. The samples after estimation of free moisture were stored at 110°C until required for water of crystallization. No further mechanical preparation was given to the sample. The water evolved by this method has been termed 'water of crystallization' but would include any water due to the oxidation of hydrocarbons. At this stage no attempt was made to determine each one separately.

APPENDIX II

The Determination of Hydrogen in 6 s.w.g. coated Al–10% Si alloy welding rods

By E. J. Millett, B.Sc.

Determinations were required of the internal gas content of the core wire, and, if possible an estimate of the quantity of hydrogen likely to be evolved by reduction of moisture absorbed on the metal surface and in adherent corrosion products.

A vacuum extraction and low-pressure gas analysis system[5] was used to carry out determinations at 570°C for the Al–Si alloys by the vacuum heating method,[6,7] and by a modified version of the tin fusion method,[8] using an extraction temperature of 700°C.

Determination of Blanks

In order to determine the internal hydrogen content of the core wire, a correction must be applied for the chemisorbed film which is picked up by a freshly machined aluminium surface.[9] This is conveniently done in the vacuum heating method by carrying out a repeat determination on a sample which has been re-machined after a preliminary de-gassing. This is not practicable with the tin fusion method which was finally adopted for this series of determinations, but the blank value can be estimated by de-gassing a specimen by heating (at a temperature approaching the melting point, preferably in the liquidus–solidus range of an alloy) and subsequently carrying out a determination by fusion on the sample after re-machining.

A number of specimens were machined from the core wire of one of the available welding rods, and determination of the internal hydrogen content and surface correction were carried out, both by vacuum heating and tin fusion. The following results were obtained in ml/100 g of sample.

Internal hydrogen by vacuum heating (uncorrected)	0·26, 0·31, 0·34 0·74, 0·74
Surface correction by vacuum heating	0·12, 0·14

The last two results for the internal gas content were obtained by heating the sample dangerously close to the eutectic temperature. One such sample was inadvertently melted and gave the result of 0·62 ml/100 g.

Determinations of the surface correction by the tin fusion method initially gave results lying between 0·4 and 0·7 ml/100 g which was reduced by 0·3 ml/100 g on the specimens de-gassed at the eutectic temperature. In order to ascertain if this was a blank error inherent in the tin fusion process, determinations were carried out on some samples of pure aluminium which had been previously thoroughly de-gassed by vacuum heating. These samples were machined to $\frac{3}{8}$, $\frac{3}{16}$, and $\frac{1}{8}$ in. dia. and the surface corrections obtained were 0·00064, 0·00056, and 0·0014 millilitres of hydrogen per square centimetre, which is in good agreement making allowance for the large surface to volume ratio of the smallest specimen, with the values 0·0002 to 0·0004 ml/cm^2 previously determined by vacuum heating and the published figures[6,7] of 0·0003 to 0·0005 ml/cm^2.

It was concluded, therefore, that very careful temperature control was necessary for the full extraction of hydrogen from the alloy samples, and increased extraction of gas was obtained by the tin fusion method. A simple correction of 0·2 ml/100 g was applied throughout the determinations of internal hydrogen content.

Determination of 'Surface' Hydrogen

Corroded samples of the core wire were obtained by stripping off the coating, using a pair of pliers to crush it, and the wires were subsequently cleaned superficially by removing adherent lumps of coating with a knife under a hand lens. Although the amount of electrode coating left adhering to the specimens could be of the order of a milligram it seemed unlikely that the amount of moisture contained in it would interfere seriously with a determination, and this was supported by the results of determination carried out by both methods.

Total hydrogen by vacuum heating 2·2, 3·0, 2·7 ml/100 g
Total hydrogen by vacuum fusion 3·0, 3·5 ml/100 g.

In view of the random variations likely in the condition of the surface this was considered to be a satisfactory agreement and the blank correction would in any case be less than the experimental error.

The contribution of hydrogen from the corroded surface is then obtained by subtracting the internal gas content from the total.

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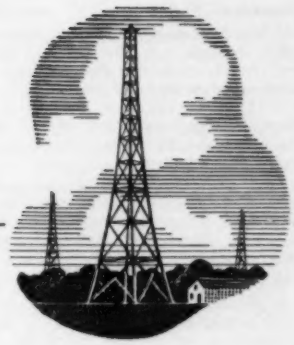
NEWS
and Announcements



News of the Institute and Branches
B.W.R.A. · Other Societies ·
and Industry

INSTITUTE OF WELDING

Annual Dinner 1958

The Annual Dinner of the Institute, which is the opening function of the Autumn Meeting, will be held at the Park Lane Hotel on Wednesday, 5th November 1958. Companies and firms which are accustomed to entertain their friends at this dinner are asked to note the date, which it is hoped will not conflict with any similar engagement in which they are likely to be interested.

BRANCH NEWS

East Midlands—Annual Meeting

The new Branch President, Mr. H. L. Palmer, B.Sc., A.M.I.Mech.E., was not able to attend the Annual General Meeting held at the Victoria Station Hotel on 8th April, through absence abroad. He will receive the insignia from the retiring President, Wing Commander J. M. Aiton, D.L., at the first meeting of the next session.

The retiring Chairman, Mr. A. R. Heaton, introduced his successor, Mr. K. H. Ewen, A.M.I.Mech.E.

At the end of the official business two films were shown. The first, a colour film "Argonaut Welding", illustrated the extreme versatility of this process, and showed the metal transfer by droplets across the arc in slow motion.

The second film, "Welding in Boiler Manufacture" reviewed the history of welding applied to boilermaking and illustrated the use of fire welding, oxy-acetylene fusion welding, electrical resistance flash welding, manual metal-arc welding, the submerged arc, and the resistance stud welding of elements on to boiler tubes.

East Midlands—Joint Meeting on Welding Research

At a joint meeting with the Lincoln Engineering Society, held on 14th March, Dr. Weck presented a paper on the "Importance of Welding Research to Industry".

Among the subjects reviewed, specific examples that were quoted included work on portal frames and their economic design, reinforcement of nozzles and openings in pressure vessels, and brittle fracture especially in relation to the materials used in the construction of nuclear power plants.

Liverpool and District—Tanker Construction

The meeting held on 10th December was a shipbuilders night, for the speaker, Mr. L. Johnson, gave a very detailed account of "Some Aspects of Welded Tanker Construction". Dealing first with design aspects Mr. Johnson gave some of the reasons for the choice of proportions and sizes of modern tankers to provide economic cargo capacities. He then outlined the considerations leading to different methods of framing for the ship's structure. In regard to welding he referred to some of the earlier failures in the details of riveted constructions and described how the problems were being overcome in the design and construction of welded connections.

Liverpool and District—Alloy Steels for Pipework

Mr. W. L. Roe gave the lecture to the Branch on 11th March, his subject being "The Welding of High-Temperature and High-Pressure Pipework". He referred to the more commonly used pipe steels and gave recommendations for limits of service temperatures, types of welded joint, and the preparation of tube ends, together with suitable welding processes and procedures. These included details of filler rods, electrodes, preheating, and post-weld heat-treatment. Non-destructive examination of the welded joints, repairs to faulty welds, and the training, testing and supervision of welders were also discussed.

London Branches Joint Dinner

There was a large attendance at the Annual Dinner of the North and South London Branches, held at the Palmerston Restaurant on 7th March. It was presided over by Sir Charles Lillicrap, K.C.B., M.B.E., President of the South London Branch and also of the Institute.

The Principal Guest was Mr. W. S. Atkins, who proposed a toast to the Institute and the London Branches. He spoke of the development of the work of the Institute in recent years and drew attention to the duty of the Branches to support this development. He considered that more should be done to encourage greater attendance at technical meetings in the branches. In his reply, Sir Charles endorsed the remarks of Mr. Atkins. He referred in particular to the success of the Institute's experiment in setting up the School of Welding Technology and to other work in the educational field.

The toast to the guests was proposed by Mr. L. C. Percival, President of the North London Branch, and the reply was given by Professor J. G. Ball, who spoke of the essential part that welding had played in the development of nuclear power.

South London—Film Show

The programme for the meeting on 13th March consisted of three British films—"The Dounreay Sphere", "The Film that Never Was", and "Craftsmen Welders". Mr. Strevens of Babcock and Wilcox Ltd., made many interesting additions to the commentary on the latter film, and there was an animated discussion. Two German films were shown in the second half of the meeting: "Arc Cutting of Aluminium", and "Welding the N.S.U. Quickly Moped". Mr. Burgess of the Aluminium Development Association commented on the two films.

South London (Medway Section)—CO_2 Shielding

Mr. D. B. Tait of Quasi-Arc Ltd. was the speaker at the meeting held at the Medway College of Technology on 19th March. He described the Fusarc/CO_2 equipment in detail and discussed some of the merits of the process together with illustrations of its applications in fabrication. Later, Mr. Sheppard demonstrated the equipment by making butt and fillet welds.

Wolverhampton—Forum

At the March meeting, held at the Holly Bush Hotel on 19th, a Welding Forum took the place of the advertised Branch member's paper, under the chairmanship of Mr. R. W. Allan. The panel members included Mr. P. J. L. Leder and Mr. G. C. Poxon of Quasi-Arc Ltd., Mr. R. J. Fowler of Horseley Bridge, and Thomas Piggott Ltd., and Mr. Colin Spencer of Bayliss Jones and Bayliss Ltd. Some of the many questions put to the panel were as follows:

"To what extent have batteries been used as a potential power source for the inert-gas metal-arc process? What is the maximum current used? and leaving economics aside, is it practicable to use the process for welding small non-ferrous parts?"

Mr. Poxon replied that batteries have not been used to any extent as potential power sources, one reason being the great loads that would have to be stored to provide the maximum current capacity, and another being the constant need for re-charging. He thought the method would not be suitable for welding small non-ferrous parts in comparison with the variety of other equipment that was available.

"What does the panel consider to be the most significant advance in the application of the technique of welding in recent years, and what action is considered to be needed to spread a knowledge of welding matters to a wider public than is being covered by the Institute's Welding School of Technology?"

The panel were somewhat divided on the first part of this question; some agreed that developments in the automatic field were most significant, whilst others put the advent of the low-hydrogen electrode as the most important step.

On the second part, all were agreed that the Institute's School was the answer to the engineer or supervisor who wished to gain a further knowledge of welding. In regard to the practical man some concern was expressed over the poor conditions in some of the technical colleges—often the equipment was not in good order, and there was sometimes a lack of co-operation that would hamper rather than help the student.

"What types of defect have a serious effect on service life? and is it more dangerous to repair a slight defect than to leave it alone?"

This question evolved from discussion on radiography and the ever increasing stringency of inspection. The panel agreed that many small defects could be ignored but that cracks, the most serious of all, should be cut out and rewelded.

The successful meeting was closed with a vote of thanks by Mr. S. H. Griffiths, seconded by Mr. A. H. Griffiths.

NEWS OF MEMBERS

A. H. Goodger

Mr. **A. H. Goodger,** M.Sc. (Tech.), A.R.I.C., who has been head of the Metallurgical Department of the National Boiler and General Insurance Co. Ltd. for many years, formally retired from full-time service on 12th April, 1958. Among his future activities Mr. Goodger is being retained in the capacity of Consultant Metallurgist by Messrs. Kennedy and Donkin, Consulting Engineers in the electrical generation and transmission fields.

Educated at The Hulme Grammar School, Manchester, Mr. Goodger graduated at the Manchester College of Technology. He joined Messrs. Galloways Ltd. in 1913 where he served in the capacities of Metallurgist, Foundry Manager, and Joint Works Manager. He has been with the National Boiler Company since 1934.

Mr. **S. G. P. de Lange** has relinquished his responsibilities as Commercial Manager of the Industrial Products Division of Philips Electrical Ltd. He has been appointed advisor to the Company and co-ordinator in all matters affecting industrial products. Mr. de Lange has been with the Philips organization for more than 30 years. He came to England in 1936.

He is succeeded by Mr. C. W. V. Davis, who was previously Manager of the Filtration Department.

Dr. **P. H. Sykes,** a Director of B.O.C. Ltd., has been elected Chairman of the Low Temperature Group of the Physical Society.

CONTRIBUTORS TO THE JOURNAL

D. F. Adams, Research Officer on metallurgical problems with the British Welding Research Association at their Research Station at Abington, was educated at High Wycombe Technical Institute and at Battersea Polytechnic, where he obtained a Higher National Certificate in Metallurgy. He joined High Duty Alloys in 1937 and was subsequently associated with the casting and hot working of light alloys.

In 1945 he took a post at the Admiralty Engineering Laboratories at West Drayton and joined Southern Forge Ltd. in 1950 to carry out work in connection with aluminium alloy extrusion and tube production. He took up his present appointment in 1952 and has since been concerned with research on the welding of light alloys. Mr. Adams was made an A.I.M. earlier this year.

N. G. Hanson, Assistant General Manager of Swan, Hunter, and Wigham Richardson Ltd., at the Wallsend-on-Tyne Dry docks, served his apprenticeship in the Wallsend Shipyard. He went through the various drawing offices and trades and was transferred to the Repair Department in 1941.

Mr. Hanson was appointed Assistant Manager in this Department in 1943, and was promoted to Works Manager in 1945. He took up his present position in 1957.

D. F. Adams

N. G. Hanson

R. Rose graduated from the Technical University of Norway at Trondheim in 1955 as a metallurgist and from 1955–57 was engaged in research concerning gases in welds at the Engineering Research Foundation (SINTEF) of the University. Mr. Rose is at present at the Joint Establishment of Nuclear Research (JENER) Lillestrom, Norway.

OTHER WELDING SOCIETIES

American Welding Society

At the 39th Annual Meeting of the American Welding Society, held in April at St. Louis, Missouri, R. David Thomas, Jr., President of the Arcos Corporation, was elected a Vice-President of the Society.

The President, C. P. Sander of the U.S. Steel Corporation, opened the meeting, and presented a number of the Society's awards, among them the Samuel Wylie Miller Memorial Medal to R. David Thomas and the J. F. Lincoln Gold Medal to Perry C. Arnold of the Chicago Bridge and Iron Co.

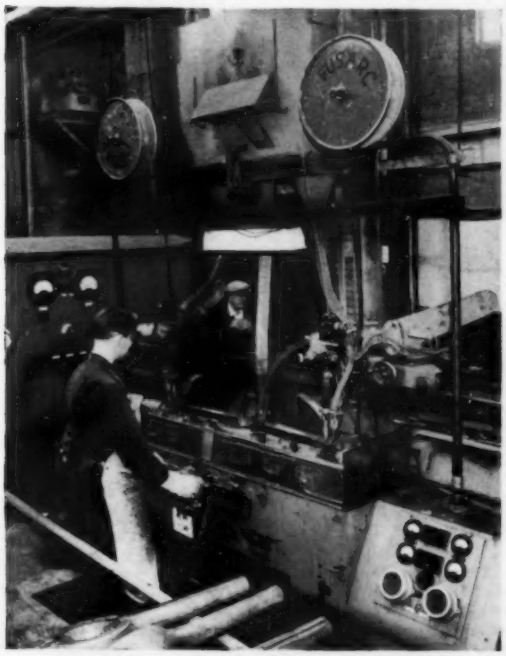
The Adams Lecture 'The Weld Heat-Affected Zone' was presented by E. F. Nippes, Professor of Metallurgical Engineering and Director of Welding Research at Rensselaer Polytechnic.

The 6th AWS Welding Show was being held in association with the Annual Meeting.

NEWS FROM INDUSTRY

Automatic Submerged-Arc Welding

Much automatic welding equipment now being produced is of specialized application but it is nonetheless of general interest, for with slight modifications any given machine may possibly be applied to the production of other components.



One example, using two submerged-arc automatic welding heads, has recently been produced by Quasi-Arc Ltd. for the welding of axle cases at the Birmingham factory of Fisher and Ludlow Ltd. The two heads can traverse independently on the split lead screw, and the pneumatic clamps forming the component jig also operate the device for retaining the flux powder in position during welding. Operation of the machine has confirmed the opinion that such automatic equipment not only reduces welding times (in this instance to one-eighth of the manual time) but materially enhances the quality of the deposit.

Pressure Welding

Another of the Joining of Metals Conferences is to be held on 19th June in the Department of Industrial Metallurgy of Birmingham University, the subject being the "Pressure Welding of Metals". The programme is as follows:

10.0—10.30 a.m.—Opening Address and Introductory Survey by Prof. E. C. Rollason (Head of Dept.).

10.30—11.30 a.m.—Industrial Practice in Cold Pressure Welding by J. A. Donelan (G.E.C.).

11.30—12.0 noon—Break for refreshments.

12.00—1.00 p.m.—Pressure Welding by Rolling by L. R. Vaidyanath, M. G. Nicholas, D. R. Milner (Dept. Ind., Met., Birmingham)

1.00—2.15—Luncheon.

2.15—3.15 p.m.—Pressure Welding by Relative Movement Between Surfaces by Dr. E. Holmes (Dept. Met., Notts.)

3.15—4.30 p.m.—Discussion, followed by Tea.

Applications for enrolment should be sent to the Secretary, Dept. of Industrial Metallurgy, University of Birmingham, Edgbaston, Birmingham 15, and marked "Pressure Welding Conference". An enrolment fee of 15*s*. will be charged; it includes the cost of luncheon and refreshments. The last date for receipt of applications is Thursday, 12th June.

Cleveland Bridge Open Day

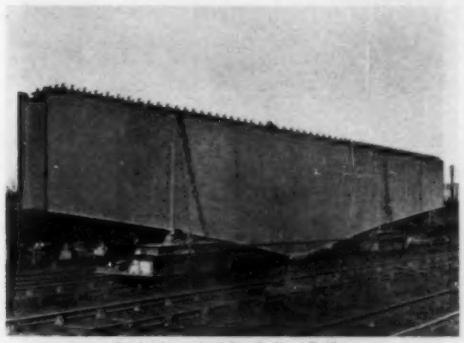
Many of the leading consulting engineers in the structural field, representatives of local and national transport authorities and of the welding organizations attended the Open Day of the Cleveland Bridge and Engineering Company at Darlington on 4th March.

The Company, famous throughout the world for the bridges, power stations, and a variety of other steel structures that it has built, had on show many examples of welded fabrications going through the shops.

For the first sixty odd years of its existence most of the work was designed for riveting. From the late 1930's, however, and particularly since 1939, an increasing use has been made of welding techniques. These have resulted in speed and economy of production by allowing reductions in the size of scantlings and so reducing the amount of preparatory work and handling. Cleveland's welding shops have made substantial contributions to the development of new welding techniques as applied to large fabrications, especially in automatic and semi-automatic processes.

Visitors saw being fabricated in the Welding Shop tower legs for a rocket test stand. These are of box construction, but to allow all welds to be radiographed corner pieces have been made from plate pressed into angle shape. The seams are welded on both sides with automatic submerged-arc equipment. Box columns of conventional type were also being submerged-arc welded for the Wilton Power Station.

Further heavy welded fabrications were seen in the West End of the Main Shop, where the main work in progress was on steelwork for Samlesbury Bridge, Preston. The parts seen were the haunch sections of the eight box girders forming the bridge.



Centre haunch of Samlesbury Bridge

The flanges included N.D.1 material up to 2 in. thick, and the butt joints are preheated to 200°C. Low-hydrogen electrodes are used for all manual welding and the joints are subsequently radiographed. There are many butt joints in the ¾-in. thick web plates; these are made with a modified submerged-arc welding head mounted on a carriage, enabling welds to be deposited in one pass at a speed of 12 in/min.

Since accuracy of assembly for site welding is imperative, the first girder made was trial erected to determine shrinkage allowances and true setting out preparatory to making jigs for setting up the later sections.

In this shop too was seen welding equipment used on other forms of construction.

Although the East End of the Main Shop is chiefly reserved for difficult riveted work, some welded fabrication was in progress; in particular, some heavy welded girders for the Margam Bessemer plant. These girders range in weight from 8 to 17 tons and include flanges made from N.D.1 and N.D.11 notch-tough steels. Special precautions are necessary in assembly and welding to counteract distortion, for the flanges are to be machined after welding.

The degree of precision attained and the proper planning exercised by the Company was shown by inspection of other important parts of the works—the steelyard, the template shop and setting-out bay, the plate preparation bay, which contains straight line and profile cutters, the erection yards and the X-ray shop. Cleveland Bridge is justly proud of its past achievements. It is certain that in the future it will make further important contributions towards even greater applications of welded structures.

Technical Films

Two new films "The Flame Hardening Process" and "Advances in Oxy-Acetylene Welding Techniques" are available on loan from the Sales Technical Service Department of British Oxygen Gases Ltd., N. Circular Road, London, N.W.2.

Welding Internal Seams in Narrow Vessels

A new piece of automatic equipment has been produced by Quasi-Arc Ltd. for welding internal seams, longitudinal or circumferential, in vessels with diameters of from 2 ft 6 in. to 10 ft, and 10 ft long. The automatic welding head is fixed to a slender boom, which can be traversed vertically through motor-driven gearing on to a lead screw. The position of the welding head is controlled by a guide wheel running in the weld groove, and there is fine adjustment to the welding nozzle.

Since the welding head is fixed horizontally the equipment is intended for use with traversing roller manipulators.

Welded Aluminium Boat

The first Fast Patrol Boat to be built in the U.K. with the complete hull—keel, frames, bulkheads, deck and side framing—of welded aluminium, was recently launched from the Anglesey Yard of Saunders–Roe.

The alloys used in the construction of H.M.S. *Dark Scout*, which is 71 ft 4 in. long, were NP5/6 plate, NS6 sheet, and NE6 extrusions, and were welded by the Argonaut shielded arc process. Most of the shell structure was prefabricated in large sections, and the longitudinals, keel, and bulkheads were also made in large units.

Expansion of Northern Aluminium

Field Marshall The Earl Alexander of Tunis, K.G., Chairman of the Northern Aluminium Co., Ltd., recently announced plans for the extension of the Rogerstone rolling mills and further modernization of the Banbury and Birmingham works.

Already since 1946 the Company has spent nearly £7 million on new plant. Now a further £10 million is to be spent over the next four years, most of it for new plant and buildings. It is intended to raise the annual capacity of the Rogerstone mills to 75,000 tons with provision for later increases to 175,000 tons.

The main reason for the scheme is to increase the range and lower the cost of the Company's products. Thus, wider and longer plate in aluminium and its medium-strength alloys will become available, and there will be increased scope for the production of the heat-treatable high-strength alloys for aircraft manufacture.

Heat Exchanger Vessels at Berkeley

The first of the 16 large heat exchangers for the Berkeley nuclear power station was erected early in April by the manufacturers, John Thompson Ltd. The vessel, which is 70 ft long, 17 ft 6 in. dia., and weighs 133 tons, was fabricated in the site boiler shop from large sections prefabricated in the company's shops at Wolverhampton. The 1½-in thick steel shell plates were joined by the submerged-arc process with the welding heads set in a new high-lift boom machine.

After welding, all seams were radiographed and the sections were furnace stress-relieved. At site the sections were welded manually, rollers being used to enable all welds to be made in the downhand position. These welds were also radiographed *in situ*, and local stress relieving was carried out by means of induction heating coils and auxiliary internal heaters.

When erection of the vessel in its final position was completed, and this itself was a major operation, the surrounding structural steelwork and internal heat-exchange tubing was completed.

Safety Features in New Preheating Torch

The "N.W.G." high-pressure torch, developed by the North Western Gas Board, has been designed primarily for use in shipbuilding and engineering works, for preheating steel plates, sections, and castings prior to welding and for other portable heating applications.

In general, the torches employed for these processes in shipbuilding works use acetylene, butane, or propane in conjunction with oxygen. In all designs of such torches in use, both the fuel gas and oxygen are under high pressure, and any leakage in the torches or connecting hoses may result in considerable quantities of gas escaping into confined spaces, such as tanks and the interior of hulls. Such occurrences can, of course, be very dangerous, and in the past have been the cause of several fatal accidents. The fact that propane and butane have little odour increases the hazards where these two gases are used. Apart from the possibility of explosion or firing of fuel gases, leakage of oxygen can give rise to an increased fire hazard, since any inflammable material such as workmen's clothing will burst into flame if ignited by a spark in an oxygen-enriched atmosphere.

The N.W.G. torch was designed specifically to reduce these hazards and incorporates the following features:

(i) High pressure air is used instead of oxygen

(ii) Town gas with its distinctive odour is used instead of propane or butane

(iii) The torch is so designed that the town gas will pass along the flexible hose to the torch only when the pressure inside the tube is below atmospheric. Consequently, in the event of any leakage on this hose, no gas can escape and the air drawn in passes harmlessly to the torch.

The principle of operation lies in the use of a high-pressure air supply to induce the gas flow through the small-bore flexible hose, and of a valve designed to open at just below atmospheric pressure. This valve will not open unless air is passing to the torch, and disconnection of any hose connections results in immediate closing of the valve.

A second protective valve on the gas supply prevents this negative pressure being communicated to any other adjacent supplies, and the combined unit satisfies the requirements of the 1948 Gas Act.

The speed of heating with this torch is naturally slower than for types using oxygen, but this is not found to be an inconvenience in practice and is more than repaid by the greatly reduced cost of fuel and air in place of oxygen, in addition to the safety features obtained.

Further details of the torch are as follows:

Portability—Weight 2¼ lb complete. Length 1 ft 6in. (variable) Connections ⅜ in B.S.P.T. male cone union to accept ⅜ in. bore hose to 30 ft.

Output—Maximum 125,000 B.Th.U per hour. Consumption 15–275 ft³ gas/hour.

Air Supply—The maximum output is obtained with air at 100 lb/in². Satisfactory results are obtained from 50 lb/in².

Construction—Of light but robust construction. The flame-retention burner nozzle may be round or rectangular, and will withstand arduous duty.

Flame Stability—Within the gas rates given above the flame is stable, and lighting back, or flame blow-off, can be obtained only by deliberate and precise valve adjustment.

The design and safety features are the subject of a Patent Application.

STUB ENDS

▶The managing director of Sciaky Electric Welding Machines Ltd., Slough, Mr. R. W. Ayres, recently started a tour of the company's agents in S. Africa, India, and Australia. Since there is a marked potential demand for resistance welding equipment in Australia it is proposed to establish a factory in Sydney.

▶An A.S.T.M. special technical publication (No. 196) contains a number of papers on fatigue properties of high-strength steels and light alloys.

▶Suffolk Iron Foundry have appointed Mr. W. E. A. Williams Sifbronze Technical Representative for London and the South Eastern areas. (100 Preston Lane, Tadworth, Surrey—Burgh Heath 6536.)

▶Taylor Woodrow (Nigeria) Ltd. have just completed the 65-mile test pipeline, 10 in. dia., for the Oloibiri oilfield in Nigeria. As the pipe joints were welded by the 'stovepipe' method 30 welders, experts in this technique, were sent out from the U.K.

▶The United Steel Companies' Review of Progress for 1957, a specially prepared report for employees, attractively presents the Group's activities and prospects. The value of the Companies' expansion and modernization programme is illustrated by the fact that whilst steel output increased by 78% during the period under review the number of employees has grown by only 25%.

▶Crompton Parkinson (Stud Welding) Ltd. have appointed Mr. T. G. Gibson as Sales Engineer for Scotland, N. Ireland, and Eire. His activities are centred on the Glasgow Office.

▶Many examples of welded fabrication were on show at the Mechanical Handling Exhibition, held at Earl's Court from 7th to 17th May.

▶The 27th International Fair in Poznan, Poland, is being held from 8th to 22nd June.

Second Annual Gregory Award

Crompton Parkinson (Stud Welding) Ltd. announce that a second cash award of $1500 (or $2000 in scholarships for any individual or school of the winner's choice) will be given for the year's most outstanding contribution in the field of semi-automatic electric arc stud welding.

The principal award will be made to the person responsible for the development of stud welding applications or studs which shall be judged most significant on the basis of:

(1) Reducing costs for industry, or

(2) Improving the appearance, serviceability and life of a product or structure, or

(3) Performing a function not possible by any other method.

A complete stud welding unit (tool, timer, and generator power source) will also be presented to the engineering school or college whose students and faculty shall have submitted the most outstanding group of entries in this competition.

Gregory Industries Inc., Lorain, Ohio, Licence Associates of Crompton Parkinson (Stud Welding) Ltd. are the sponsors of this competition, and the awards will be presented at the 1958 Awards Luncheon of the American Society for Metals in Cleveland, Tuesday, 28th October, 1958.

Entry forms and rules may be obtained from Crompton Parkinson (Stud Welding) Ltd., 1/3 Brixton Road, London, S.W.9. The closing date for entries is midnight, 10th July, 1958.

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Engineers in the Trafford Park area require a **Welding Engineer** who must have practical experience in gas and electrical welding up to Grade *A* standard and must be also familiar with argon and submerged-arc processes. Some metallurgical knowledge is desirable but not essential. This is a superannuable staff position with excellent prospects. Applicants must give full particulars and mark applications "For the attention of the Personnel Manager", Box. No. 219.

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Metallurgist required to fill a research post in South London related to welding. Applicants should have an Honours degree or equivalent qualification, some experience of fabrication or of smelting. Commencing salary £950. Box No. 221.

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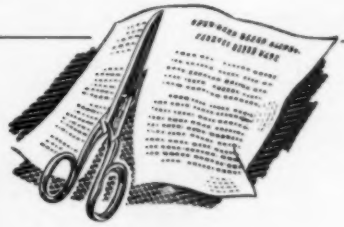

CONTENTS OF PERIODICALS RECEIVED

THIS SECTION is intended to give a survey of the current welding literature received by the Institute of Welding Library. The contents lists are not exhaustive; only the main articles in welding journals are listed, and reprints from other journals and short notes are generally excluded. In addition, welding articles from other periodicals are listed. Abstracts of welding literature are given in the *Bibliographical Bulletin of Welding and Allied Processes*, published by the International Institute of Welding, and details of this may be obtained from the Secretary of the Institute of Welding.

Industry and Welding (U.S.A.), 1958, vol. 31, No. 3, March

Special report on Weld Radiography
Part 1: Weld radiography in the shop. (28, 62–63.)
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Part 5: Comparison chart of portable X-ray equipment. (36–39.)
Part 6: Comparison chart of stationary X-ray equipment. (40–41.)
Part 7: Selecting film for weld radiography. (42–43, 91.)
Part 8: Radiography and safety. (44–45.)
How Martin Seam welds honeycomb cores. (50–53.)
Joint design for titanium missile applications: Part 2, Conclusion, B. L. Baird and C. W. Handova. (56–58.)

Przeglad Spawalnictwa (Poland), 1958, vol. 10, March

Characteristic properties of basic electrodes, J. Wegrzyn.
Submerged-arc hardfacing of mill rolls, K. Sniegoń.
PA-2 boiler damage in a fishing trawler, M. Myśliwiec.

Schweissen und Schneiden (Germany), 1958, vol. 10, March

The influence of the welding conditions on the composition and on the structural constitution of deposited Chrom-hard-metal-alloys, Dr. Helmut Koch and Dr. Elmar Bernholz. (71–76.)
The efficiency of current supply systems for electric-arc welding, Dr. E. Bergmann. (77–79.)
Welding of titanium, Dr. Klaus Rüdinger. (79–86.)
Welding of spheroidal graphitic cast iron, Dr. Jürgen Ruge and Werner Zitzelsberger. (86–90.)
Tensile strength and bending angle for fusion welded notched specimens, H. Fiehn and Dr. K. Tske. (90–93.)

Welding and Metal Fabrication, 1958, vol. 26, No. 4, April

The new electric melting shop at Brymbo steel works, W. H. Diamond and S. J. Hinder. (120–125.)
Selenium rectifiers for electric welding, L. H. Morgan. (126–133.)
Electro-slag welding in the production of boiler shells, V. I. Rabinovich. (134–137, 156.)
High energy forming, W. Edwards. (138–139.)
Automatic welding on the trans-Canada pipe-line. (140–141.)
Improved production of aluminium plate: Northern Aluminium Company's expansion at Rogerstone. (142–144.)
Fabricating copper alloy trunking, W. G. Brown. (145.)
Construction of railway rolling stock: extensive use of semi-automatic welding. (146–149.)

Zvaranie (Czechoslovakia), 1958, vol. 7, No. 3, March

Electric gas-shielded welding and the economics of its application, B. E. Paton. (65–69.)
Cold press welding of grey cast iron, R. Krnak. (69–73.)
The question of welding grey cast iron, L. Muncner. (73–74.)
Automatic resistance butt welding from the view-point of economy in production, B. Vrana. (75–79.)
Interchangeable time regulating panels for asynchronic control of spotwelding machines, Karel Jarsky. (79–83.)
Starting and development of welding in the Leninworks in Pizen, Tomas Unger. (84–87.)

Other Journals

How to weld copper and its alloys, part 2, L. F. Spencer. (*Steel* (U.S.A.), 1958, vol. 2, 24 February, pp. 90, 93, 96, 98.)
Three new pore-free ways to weld aluminium, F. R. Baysinger. (*Iron Age* (U.S.A.), 1958, vol. 181, 13 March, pp. 112–115.)
Some metallurgical aspects of welding higher strength steels, D. Birchon. (*Machinery Lloyd*, 1958, vol. 30, 29 March, pp. 71–73.)
Bonding magnesium: Chance Vought developments for airframe and missile applications, B. A. Forcht. (*Aircraft Production*, 1958, vol. 20, April, pp. 131–140.)
Welding in South Africa, F. J. A. Van Reenen and V. J. D. Hill. (*Engineer and Foundryman* (S. Africa), 1958, vol. 22, February.)

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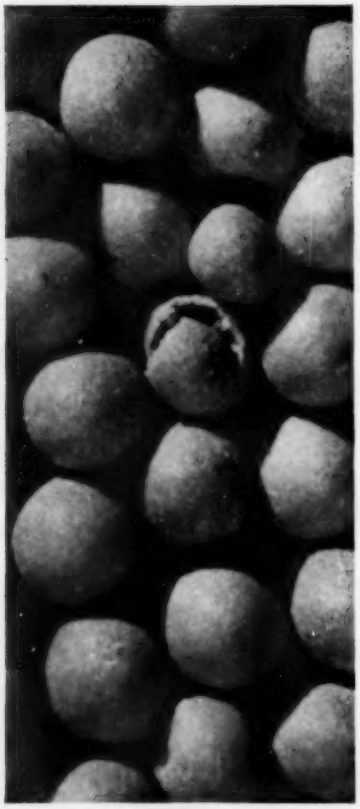

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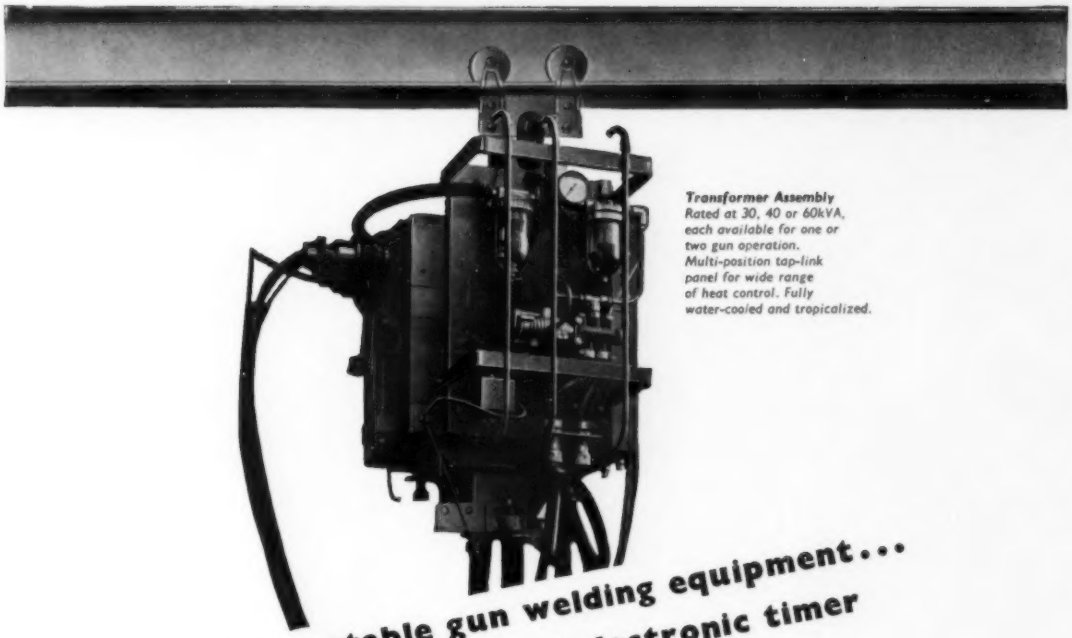

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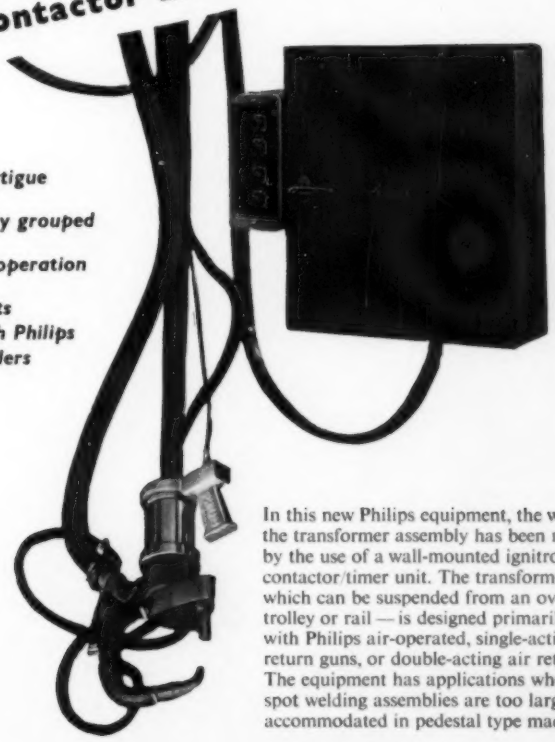

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